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# HOT DUCTILITY OF DIRECTLY CAST STEELS WITH DIFFERENT CARBON CONTENTS

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## ABSTRACT

Tensile samples of Nb containing steels with different carbon contents have been cast in situ, cooled to test temperatures in the range 800-1000 °C and tested at a strain rate of $3x10^{-3}$ $s^{-1}$.

The austenite grain sizes of the melted and resoldified samples showed real dependency on the carbon content of the steel. This resulted in variation in hot ductility values obtained for the steels examined, as increasing the carbon level from 0.014% through 0.1% to 0.14% has deteriorated the hot ductility of the steel due to grain coarsening, but a further increase to 0.18% C produced improvement due to grain refining.

The effect of the carbon content of the steel on the austenite grain size has been explained in terms of the peritectic reaction.

## INTRODUCTION

Practical observations have shown that high temperature cracking susceptibility of micro-alloyed steels during continuous casting process (CCP), depends largely on carbon content (1) as illustrated in Fig. 1. Thus, many investigations have been carried out to understand the role of carbon in the ductility behaviour of micro-alloyed steels (2-6).

The researchers simulated the CCP with hot tensile test, in which the steel, after processing, is reheated above the solution temperature of the micro-alloying precipitates (~1330 °C) and then cooled to test temperatures in the range 700-1100 °C, where cracks have been found to be very common. Cooling conditions were chosen to be similar to those encountered during CCP and the strain rate for tensile testing is chosen to be approximately equal to that undergone at the surface of the strand when it is straightened.

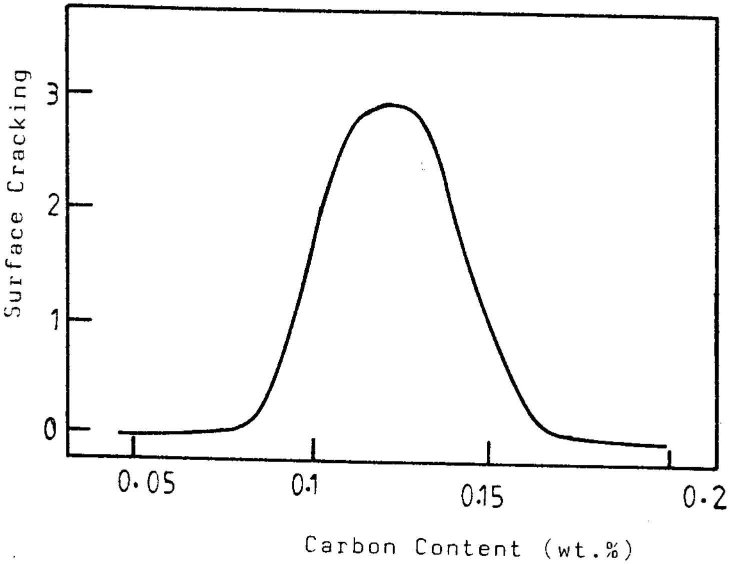


**Fig. 1: Schematic illustration showing the effect of carbon on surface cracking of continuously cast slabs (after ref. 1)**

These studies reported mixed views on the effect of carbon on hot ductility of micro-alloyed steels. While Ouchi and Matusmoto (2) and Maehara and Ohmori (3) suggested that increasing the carbon in the range 0.05 to 0.3% has no effect on hot ductility of Nb containing steels (C-Mn-Al-Nb), Hannerz (4) and Suzuki et al. (5) found an improvement in hot ductility on raising the carbon content in a similar carbon range for Al containing steels (C-Mn-Al).

Recently, Mintz and Mohamed (6) examined the hot ductility of C-Mn-Al and C-Mn-Al-Nb steels with carbon content in the range 0.05-0.15% and 0.014-0.16% respectively. They found that hot ductility of the Al containing steels is insensitive to the carbon level, but there was little influence of carbon on hot ductility of Nb containing steels.

They attributed the better hot ductility associated with low carbon steel compared to the higher carbon steels in terms of lower volume fraction of NbCN precipitation (Small particles containing niobium, carbon and nitrogen) present at the austenite grain boundaries and within the grains.

The NbCN precipitation at the grain boundaries is effective in preventing grain boundary mobility at high temperatures, leading to poor ductility and intergranular failure (7,8). The matrix precipitation is effective as well in reducing hot ductility. It has been suggested that the combination of the extensive matrix precipitation of NbCN and the precipitate free zones observed sometimes adjacent to the grain boundaries, concentrates stress at the austenite grain boundaries and hence promotes intergranular fracture and low ductility (3, 9-11).

It appears from the above information, that research results obtained from the hot tensile tests of the reheated specimens counteract the practical observations noted by Gray et al. (1) . However, the conditions of reheated samples are very different from those experienced during CCP, which show generally coarser micro-structure, pronounced segregation of impurities to the interdendritic boundaries ... etc. The aim of this work is to examine the ductility of Nb containing steels with different carbon contents after melting the samples "in situ" to be closer to the conditions of CCP.

## EXPERIMENTAL

Four types of steels chosen for examination were supplied in the hot rolled state, with finish rolling at 1000 °C and 12 mm thick plates. The full composition in percent by weight are given in Table 1, which shows similar compositions except the carbon content.

**Table 1: Composition of the Steels Examined (wt. %)**

| Code | C | Si | Mn | P | S | Nb | Al | N |
|---|---|---|---|---|---|---|---|---|
| A | 0.014 | 0.36 | 1.48 | 0.015 | 0.006 | 0.028 | 0.033 | 0.007 |
| B | 0.1 | 0.42 | 1.39 | 0.01 | 0.007 | 0.026 | 0.036 | 0.0075 |
| C | 0.14 | 0.4 | 1.4 | 0.009 | 0.007 | 0.03 | 0.033 | 0.007 |
| D | 0.18 | 0.4 | 1.43 | 0.01 | 0.0066 | 0.028 | 0.032 | 0.0075 |

Tensile samples having a length of 70 mm and diameter of 7.93 mm were machined from the plates with their axis parallel to the rolling direction. A 2 mm diameter hole was drilled from one end of each sample to the mid-length of the sample so that a thermocouple could be inserted. Samples were placed inside silica tubes with 0.2 mm diametrical clearance, then heated by an induction heating unit, described elsewhere (12), so that approximately 22 mm of the length at the mid-length position could be melted. The molten region was contained in the tolerance fitted silica sheath which in turn was surrounded by a further silica tube, through which Argon could be circulated to prevent oxidation, Fig. 2.

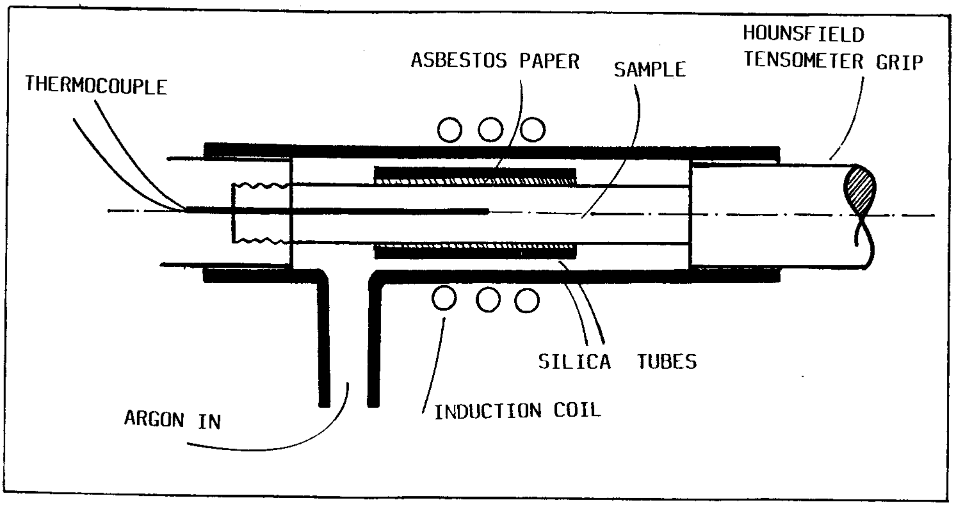


**Fig. 2: Experimental arrangement for measuring hot ductility in "as -cast" conditions**

Samples were melted at 1560 °C for 5 minutes, resolidified and cooled at 60 °C per minute to the required test temperature in the range 800-1000oC . They were then held for five minutes at test temperature before being strained to failure on a Hounsfield Tensometer using a strain rate of $3x10^{-3}$ $s^{-1}$. After fracture, the Argon gas flow was increased to cool the samples rapidly.

It should be noted that one sample from each type of steel was quenched with Argon immediately after solidification and cooling to 1100 °C to enable measuring the initial grain size at 1100 °C prior to straining at the desired test temperatures, using the linear intercept method on transverse sections.

The 5-minute holding time at melting and test temperatures, aimed homogenizing treatment. The chosen values of cooling— and strain rates were simulating the values experienced during the CCP (10, 13). The values of test temperatures have been selected because this is the temperature range at which minimum ductility occurs (6) for steels with different carbon contents.

Carbon extraction replicas were taken close to the fracture surfaces for test temperatures of 900 °C and examined using a JEOL 100 KV TEM, operating at 60 KV. Fracture surfaces were examined using a JEOL T100 SEM, operating at 25 KV.

## RESULTS

The hot ductility curves of Reduction in Area against test temperature for the four steels examined in the present work are shown in Fig. 3. Generally, all steels showed poor ductility at test temperatures of 800 and 900 °C. One of the steels which contained 0.14% C showed this behaviour even at test temperature of 1000 °C, at which the ductility of other steels started to recover. It is also evident from

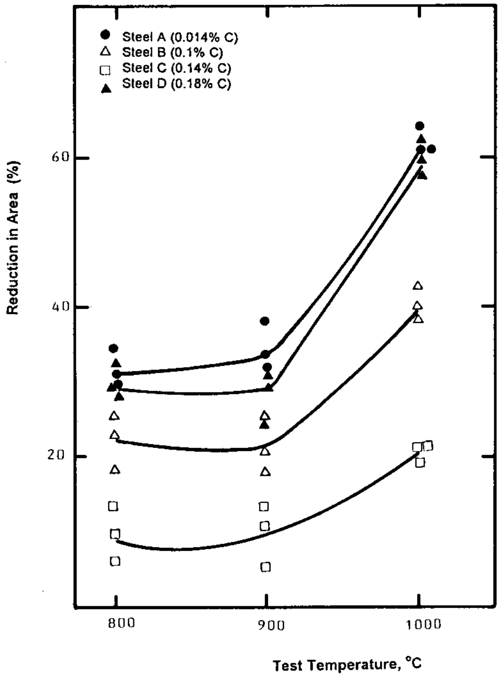


Fig. 3: Hot ductility curves for the examined steels

the figure that increasing the carbon from 0.014% through 0.1% to 0.14% has deteriorated the ductility (codes A, B and C) from ~30% to ~10% at 800 and 900 °C, and from ~60% to ~20% at 1000 °C. However, increasing the carbon content further from 0.14% to 0.18% (codes C and D), improved the hot ductility to the extent that reduction in area values for steel D (0.18% C) which contains the highest level of carbon, were very similar to steel A which contains the lowest level of carbon (0.014% C).

The improved ductility at 1000 °C for steels A and D (R of A ~60%) containing the lowest and highest levels of carbon respectively, is attributed to the occurrence of dynamic recrystallization (DR), during the test, as it was possible to detect it from the form of the load-elongation curves, Fig. 4. The fall in the rate of strain hardening leading to the peak in stress followed by work softening and fluctuations is evident of DR (14). This has been observed at test temperature of 1000 °C for steels A and D as can be seen in Fig. 4.

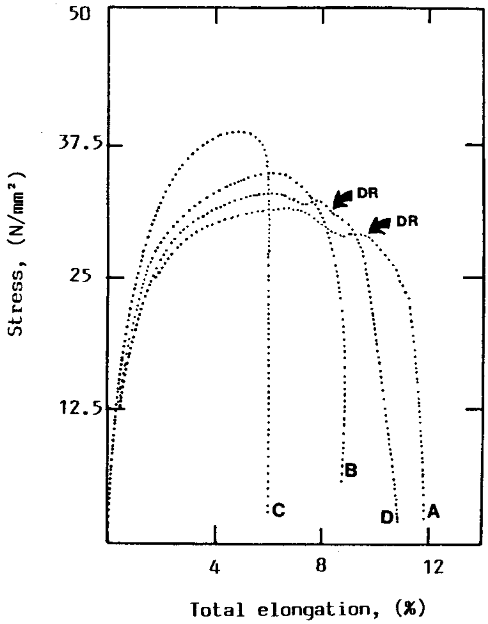


Fig. 4: Stress-Total elongation curves for the four steels examined at 1000°C

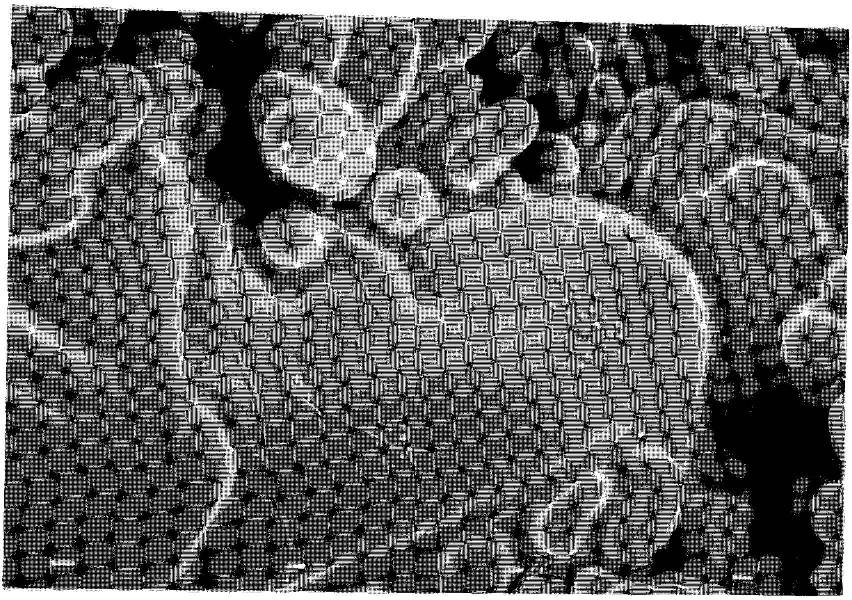
# MICROSCOPY

Samples with low ductility values ( < 40 %) were examined under the SEM. The failure mode was a mixture of interdendritic and intergranular, as shown in Figs. 5-6. The intergranular failure was in turn a mixture of intergranular microvoid coalescence and intergranular decohesion. The former is evidenced by the ductile dimples on the fracture surfaces and the latter by the flat featureless facets. Mintz and Mohamed (15) suggested that the interdendritic part of the fracture is probably where micro-shrinkage has occurred, and once 'necking' starts to take place, the fracture path will seek such regions.

Examination of carbon extraction replicas under the TEM, for samples tested at 900 °C, showed fine precipitation of NbCN (10-30 nm dia.) at the austenite grain boundaries and within the grains, Figs. 7-9. In some cases precipitate free zones were observed adjacent to the grain boundaries, Fig. 9. The precipitation was sparse in the 0.014% C steel , Fig. 7 and extensive in the other steels (codes B, C and D), Fig. 8. This has been expected from the solubility data, where raising the carbon content of the steel would increase the amount of NbCN precipitation (6). It should be noted, that the degree of the extensive precipitation in steels B, C and D did not vary significantly, which may reflect the small increments of the carbon content in the steels.

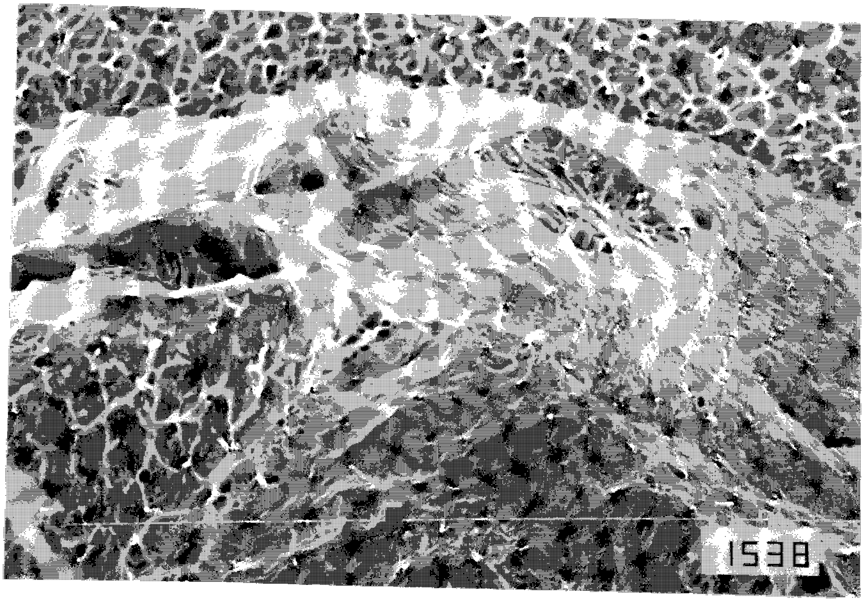
Coarse NbCN eutectics were also observed. They were similar to those shown by Mintz et al. (16) . In addition, MnS were observed in the steel samples examined with similar precipitation pattern and they contained some iron in the range of 6%-12%, Fig 10. Mintz and Mohamed (17) have shown that sulphides can deteriorate hot ductility of micro-alloyed steels, but such an explanation can not be used in the present instance as steels examined in the present work contained similar amounts of Mn and S.

The austenite grain boundaries were revealed by chemical etching in 5% picric acid solution, Fig. 11. The initial austenite grain size at 1100 °C, as a function of carbon content for all the steels examined is given in Fig. 12. In general the initial grain sizes were very coarse (670-102 µm), which is not surprising for "as -cast" steels . It can be seen also that the grain size is proportional to the carbon content up to 0.14%. This behaviour has not been changed by increasing the carbon content further to 0.18% (codes C and D), as the grain sizes for 0.14% and 0.18% C were 1020 and 720 µm respectively.



(SEMX500)

**Fig. 5: Interdendritic type of fracture observed in steel C (0.14% C) after testing at 900°C, showing coarse angular Mn (Fe) S inclusions**



(SEMX500)

**Fig. 6: Example of a mixture of intergranular Micro-void coalesence and intergranular decohesion, observed in steel C (0.14% C) after testing at 900°C**

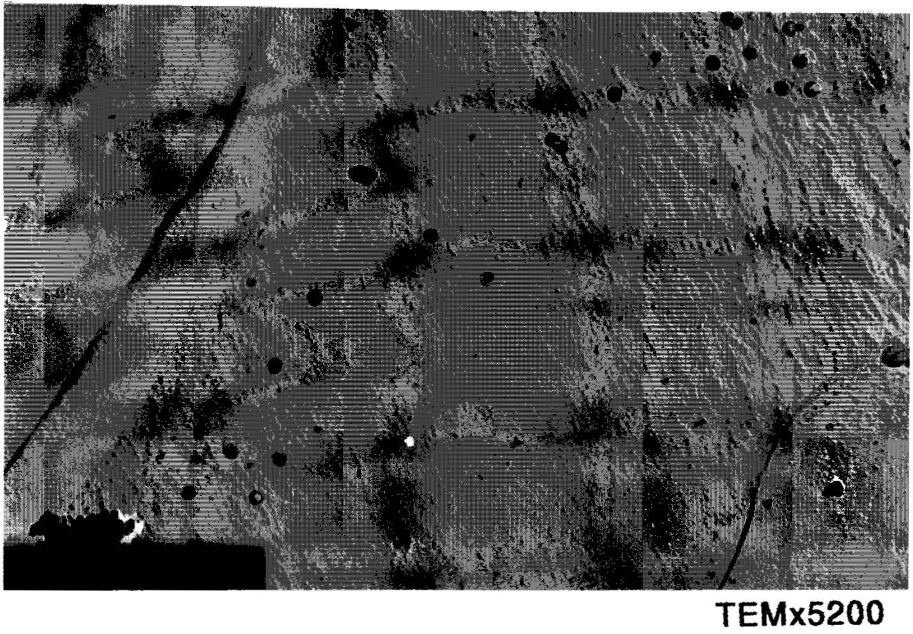


**Fig. 7: Example of NbCN precipitation observed at test temperature of 900°C, in steel A**

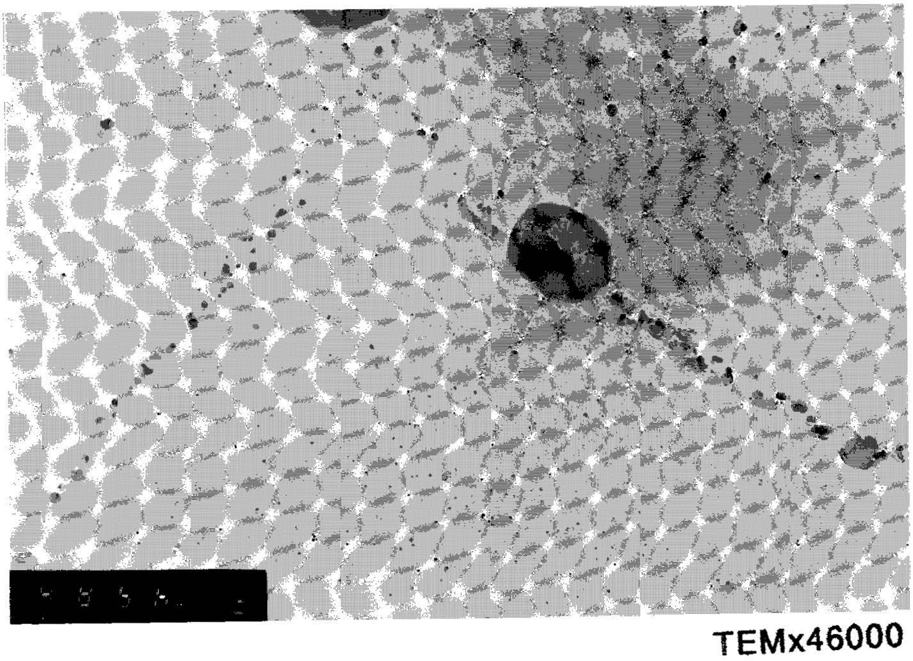
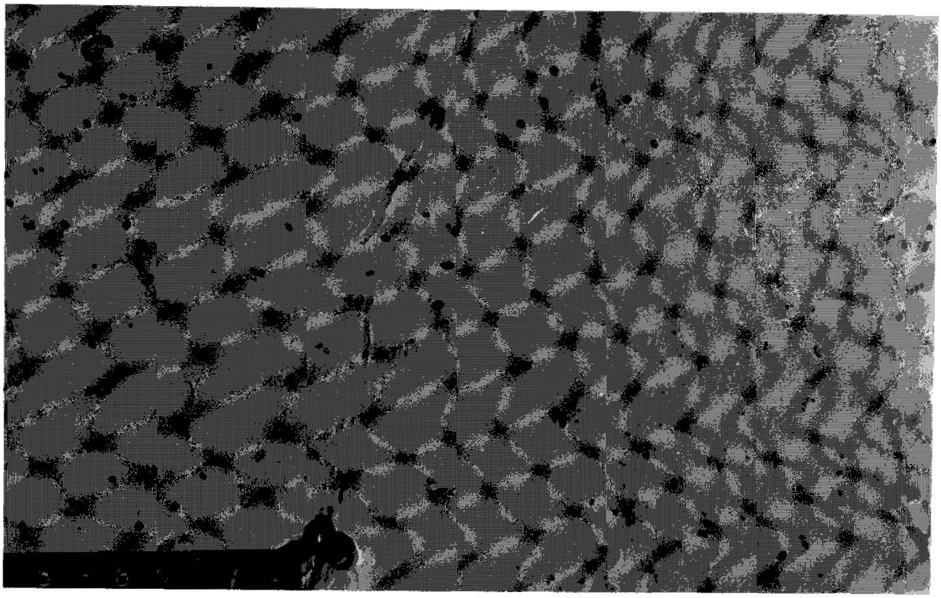
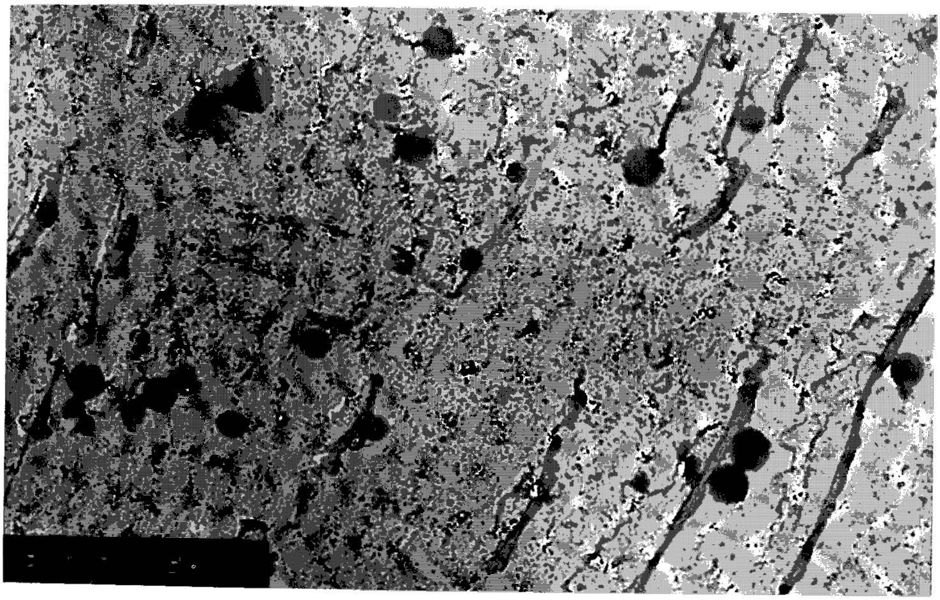


**Fig. 8: Precipitation of NbCN observed at the austenite grain boundaries in steel C at test temperature of 900°C**



TEMx45000

**Fig. 9: Example of NbCN precipitation in the matrix and the precipitate free zones in steel observed in steel C at test temperature of 900°C**



TEMx500

**Fig. 10: Example of MnS inclusion containing some iron, observed in steel B at test temperature of 900°C**

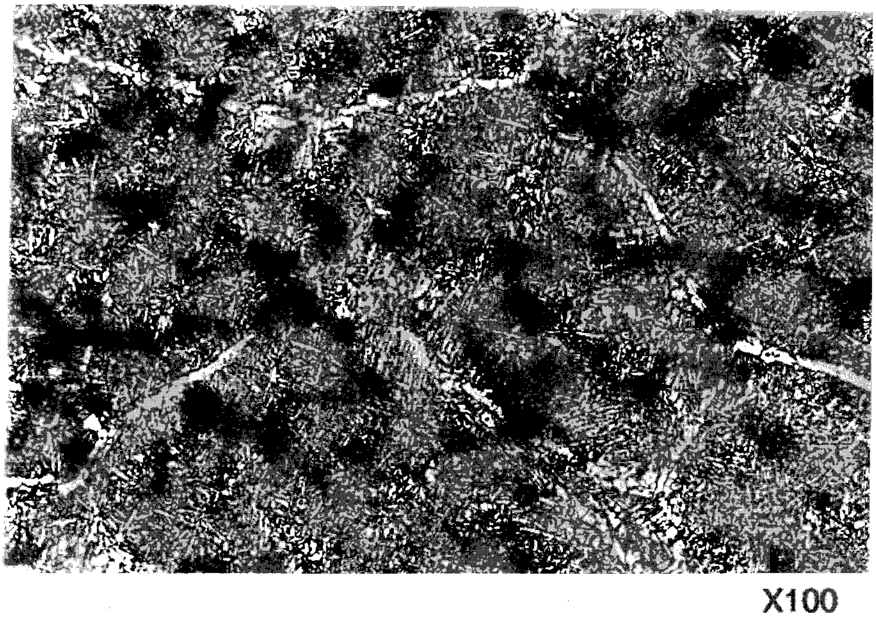


**Fig. 11: Coarse austenite grains obtained at 1100°C for steel B**

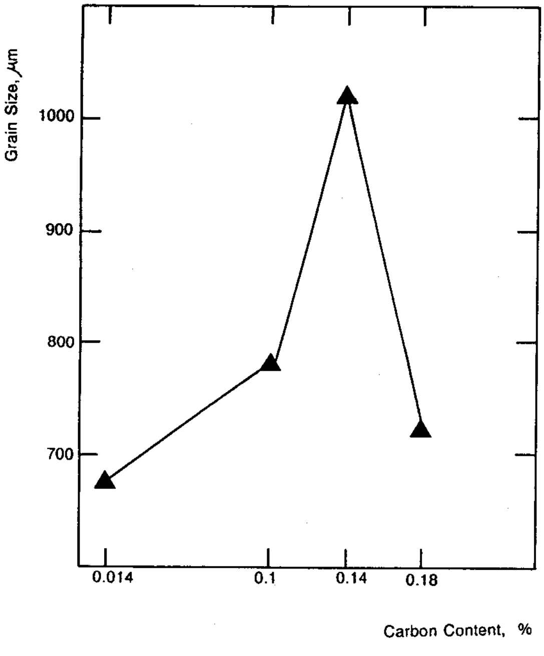


**Fig. 12: Variation of the austenite grain size with carbon content at 1100°C**

## DISCUSSION

As has been stated in the introduction to this work, tensile samples in the reheated condition have shown that ductility of the Nb containing steel deteriorates when its carbon level was raised (6). This influence of carbon content has been attributed to the increase of NbCN precipitation at the austenite grain boundaries and within the grains, which indicates that NbCN precipitation is the dominating factor controlling the hot ductility of steels. In the present work , the hot ductility of the Nb steel in the as-cast state, showed different behaviour. Steel D which contains the highest level of carbon (0.18 %) although having the maximum amount of NbCN precipitation, has better hot ductility than steel B and C, containing 0.1% and 0.14% carbon respectively. This indicates that the hot ductility, in the as -cast state is not governed by the NbCN precipitation alone. It seems to be that the hot ductility in this case is controlled by the initial austenite grain size of the steels and the effect of precipitation is of secondary importance.

The initial austenite grain sizes of the steels after solidification and before cooling to test temperatures were closely examined at 1100 °C, Fig. 12. The measurements showed that the austenite grain size varied significantly with carbon content, and hot ductility is related to the grain size (compare Figs. 3 and 12), i.e. the finer the grains, the better the hot ductility.

Many researchers (10, 18-20) have proved that coarsening the grain size would reduce the hot ductility of metals. The reason for coarse grain structures having poor ductility has been explained by Evans (21) as being due to the increased sliding rate producing an increased grain boundary cavity growth rate. However, Taplin et al. (22), believed that creep ductility, when intergranular fracture occurs, is controlled by the final stages of fracture rather than the nucleation and early stages of crack growth. A propagating grain boundary crack must grow through triple points, and it seems likely that the ease of propagation of a crack through triple point, and the number of triple points encountered, will be an important factor in determining creep ductility. Thus a coarse grained material, with few triple points will be more susceptible to intergranular cracking.

The interesting result in this work is the relation between the austenite grain size and the carbon content of the steel,. Fig. 12. The grain size increased with increasing carbon content up to the 0.14%, but decreased by increasing the carbon content further to 0.18%. This has not been observed in the previous work (6) for the reheated conditions, as steels with different carbon contents gave similar grain sizes.

It is suggested that such behaviour is probably reflecting the peritectic reaction of the steel. Thus, it would be expected that the maximum initial grain size would occur for the peritectic point. Above and below this point, austenite grain growth will be retarded due to the second phase accompanying the austenite. The second phase could be ferrite ($\delta$) or liquid depending on the carbon content of the steel. These phases should be transferred to austenite before the latter is being able to grow.

## CONCLUSIONS

1- Hot ductility troughs of similar width were obtained for all steels examined but the depth varied with carbon content.

2- Austenite grain growth of the as- cast steels depends on their carbon contents, being maximum for the 0.14%. Above and below this value, grain growth has been retarded due to pinnig effect of second phase in accord with peritectic reaction.

3- Hot ductility values measured as reduction in area showed real dependency on the initial austenite grain sizes of steels, the finer the grains, the better the hot ductility.

## ACKNOWLEDGMENT

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