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FRACTURE TOUGHNESS OF HIGH-CHROMIUM WHITE CAST IRON IN RELATION TO THE PRIMARY CARBIDE MORPHOLOGY

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ABSTRACT

High-chromium white cast iron typically has a very low (dynamic) fracture toughness. Since this hampers the application of this wear resistant material, it is important to find ways to introduce a certain amount of ductility. To this aim several different routes can be pursued. In the present work the focus is on the role of the morphology of the primary chromium carbides. The effect on toughness is investigated of the presence of a columnar or an equi-axed structure. Furthermore, an attempt is made to improve toughness by hot rolling the material. A key condition for this research is to have an experimental technique capable of measuring the extreme low fracture toughness. Therefore, in the work presented here the development of this technique is described in some detail.

INTRODUCTION

High-chromium white cast iron is a material consisting of a network of chromium carbides embedded in a martensitic or austenitic matrix. In fact the material can be regarded as a composite. Although the carbides solidify in the form of either an eutectic structure or as a primary phase for hypo-eutectic or hyper-eutectic material respectively, in this work they will be designated as being primary to distinguish them from carbides formed at lower temperatures, e.g. from the austenite phase.

White cast iron with a high chromium content is generally used because of its excellent wear resistance. This property can be attributed to the presence of the chromium carbides. In practice, however, application of this material is often hampered by a very low fracture toughness, both static and dynamic. The properties of the matrix and the presence of microscopic and macroscopic pores play a role in this, but again the presence of primary carbides is believed to be the main cause. The primary carbides form a network, which probably provides a low energy route for cracks to extend.

It is generally accepted that the network of primary carbides formed during solidification cannot be significantly changed afterwards [1]. On the other hand, the matrix can be changed by heat treatments. An important phenomenon is that at high temperatures there will also be some exchange of alloying elements, e.g. carbon and chromium, between the carbide and the matrix [2, 3].

To improve fracture toughness approaches are feasible which are related to the chemical composition, the casting and solidification conditions or the type of heat treatment. Alternatively, it might be possible to enhance fracture toughness by somehow breaking-up the carbide network. Since heat treatments are not very effective in doing so, the only way to achieve this, while remaining in the solid state, is to apply a certain amount of plastic deformation to the material [1, 4].

The main objectives of the research currently performed at Delft University of Technology on this subject are:

- to develop an experimental technique capable of determining the extremely low (dynamic) fracture toughness of high-chromium white cast iron. The technique must be easy to use and must lead to sufficiently discriminative and reproducible results.
- to gain insight in the relation between the morphology of the primary carbides and the (dynamic) fracture toughness. Important aspects of the carbide morphology are the orientation relative to the load, the shape and the extent to which a network is formed.

In the work presented here toughness measurements are described on a single type of high-chromium white cast iron, i.e. the volume fraction of primary carbides is more or less constant. In the first part of the paper, which is about the development of the test, the experimental obstacles are described together with the considerations which finally led to the set-up used. The second part deals with an aspect of the effect of the carbide morphology on fracture toughness, i.e. the presence of either a columnar or an equi-axed structure. Furthermore, the results of an investigation with a tentative nature are described in which slabs of cast iron are rolled at a relatively high temperature to different reduction percentages. This work is an attempt to improve the fracture toughness by breaking-up the network of primary carbides.

MATERIAL

Table 1 summarises the approximate chemical composition of the high-chromium white cast iron investigated. This composition represents a hypo-eutectic alloy. The material was cast in ingots of which the dimensions, i.e. length x width x height, are approximate 250 x (80-89) x 125 mm. Using electric discharge machining, slabs were taken from the middle of these ingots normal to the length direction with a thickness of 10, or for the purpose of the hot rolling experiments of 12.5 to 13 mm. Figure 1 shows the cast macrostructure of such a slab.

TABLE 1
CHEMICAL COMPOSITION OF THE WHITE CAST IRON [WEIGHT %]

C	Cr	Si	Ni	Mo	Mn
2.8	17	**1.8**	1.4	0.35	0.8

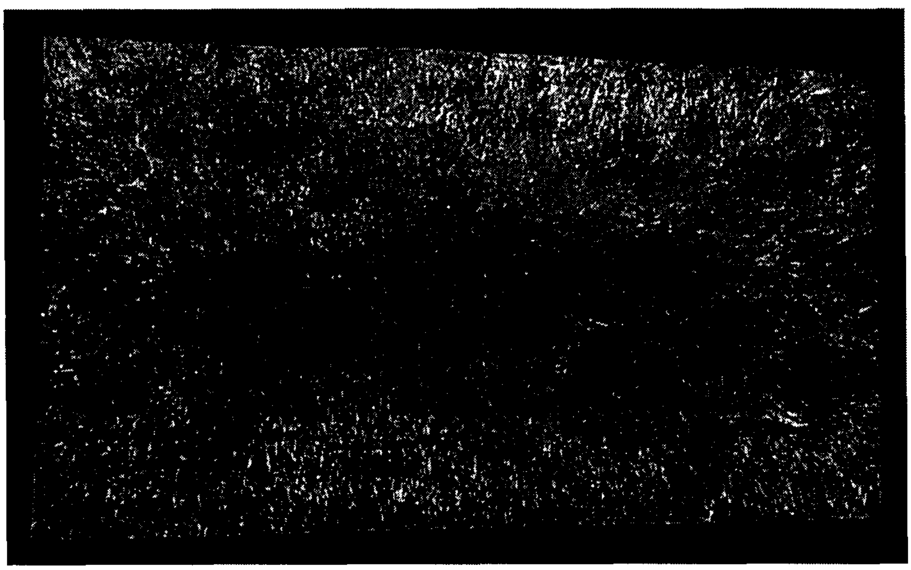

Figure 1: Macrostructure of a slab (125 x (80-89) mm) taken from an ingot showing the columnar zone in the outer region and the equi-axed area in the centre

From Figure 1 a distinction can be seen between the outer region, in which a columnar structure is present with an orientation normal to the surface, and the centre where no obvious orientation is visible, a structure which is known as equi-axed. In Figure 2 the microstructure of these two regions are shown.

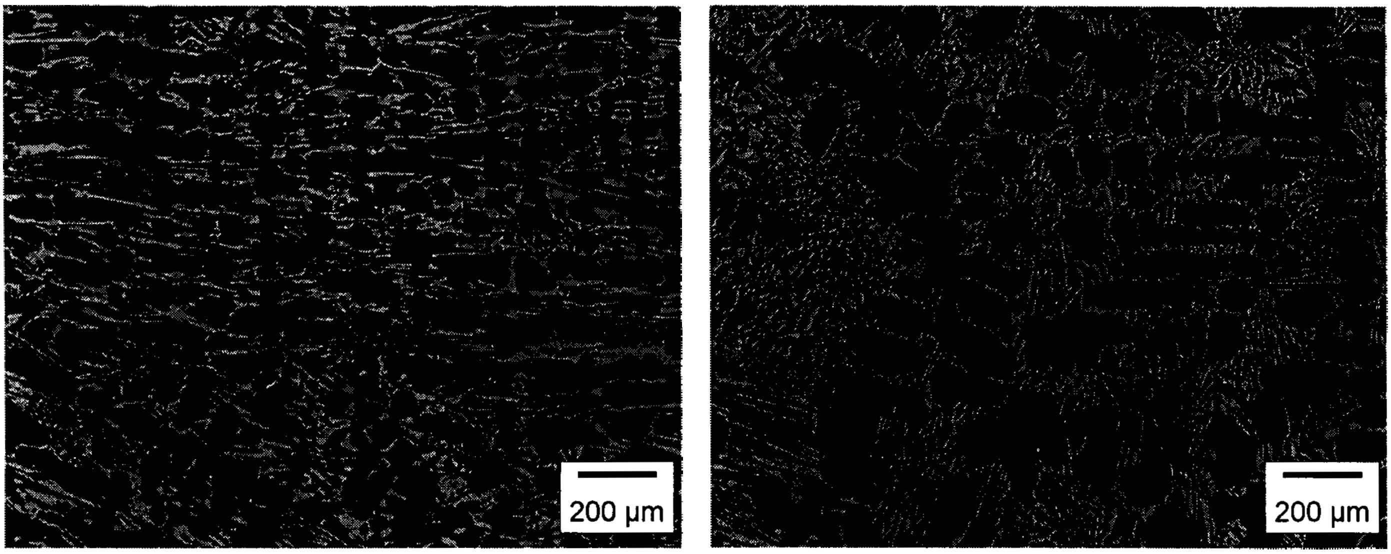

Figure 2: Micrographs of the columnar (left) and the equi-axed (right) structure in the white cast iron

MEASURING TOUGHNESS OF WHITE CAST IRON

Principle of Test Set-up

White cast iron, and especially when a high volume fraction of chromium carbides are present, does not show plastic deformation when mechanically tested. Therefore from a fracture mechanics point of view the most straightforward parameter to characterise fracture toughness is the critical (plane strain) stress intensity factor, K_{Ic}, or alternatively the dynamic fracture toughness, K_{Id} [5]. **A** problem however in using K_{Ic} is the relatively complex procedure required for measuring this parameter: specimen manufacturing, creating a pre-fatigue crack and testing itself are time consuming and expensive.

A test method which is much easier to use is the Charpy impact test [6]. In this method a notched bar with a standard size of 10x10x55 mm is dynamically loaded in 3-point bending. The idea behind this test is that by creating a certain amount of triaxiality at the notch tip and by using a high rate of loading, the conditions for the occurrence of brittle fracture are optimised and therefore the test results will be conservative. In principle from Charpy impact test the energy needed to fracture the specimen is determined and used as a toughness measure.

The experiments performed in this research are based on the Charpy impact test. An instrumented drop-weight impact tower is used with a mass of approximately 70 kg. The set-up allows the load to be digitally monitored as a function of time. In order to calculate energy values, the load versus time record is converted to load versus displacement data, a conversion which is performed by using the velocity during the test. In turn this velocity is based on the initial impact velocity and by considering the forces acting on the drop weight, i.e. gravity and the specimen load. Note that for the brittle material tested here and the relatively high impact energy of the drop weight, the velocity during the test will typically only drop slightly below the initial velocity. The energy is calculated by integrating the area under the load-displacement curve. To account for small differences in specimen dimensions the energies are divided by the area of the net cross section.

Fracture Mechanics Approach

There are two main reasons why Charpy impact test results cannot be simply correlated to a fracture mechanics parameter such as K_{Ic}:

1. For materials that exhibit a distinct amount of plastic deformation a plane strain condition at the notch tip is not guaranteed.
2. The workmanship with which the notch is machined will vary and also the radius should be small enough to simulate a natural crack.

The first problem will not play a role for the extremely brittle material under consideration here, i.e. in all cases the stress state will be one of plane strain. The second problem is often solved by introducing a fatigue crack in the specimen. This however is cumbersome in the brittle white cast iron and also contrary to the objective in this research, i.e. the development of an easy test method. Therefore the notch geometry applied here differs from those defined in the ASTM standard (V or U notch). The notch is introduced by electric discharge machining using a wire diameter of 0.3 mm and thus creating a notch tip radius of somewhere between 0.15 and 0.2 mm. It is assumed that this notch geometry sufficiently approaches a real crack so that K_I can be calculated. This assumption seems reasonable in view of the coarseness of the carbide structure, having an average spacing between the eutectic structures of the order of 200 pm (see Figure 2).

K_I is now calculated from the load $\boldsymbol{P}$ according to [5]:

$$K_I = \frac{PS}{BW^{3/2}} \cdot \frac{3\sqrt{\frac{a}{W}}\left[1.99 - \frac{a}{W}\left(1 - \frac{a}{W}\right)\left\{2.15 - 3.93\left(\frac{a}{W}\right) + 2.7\left(\frac{a}{W}\right)^2\right\}\right]}{2\left(1 + 2\frac{a}{W}\right)\left(1 - \frac{a}{W}\right)^{3/2}} \qquad (1)$$

where S = specimen span
B = specimen thickness
W = specimen height
a = notch length

In principle, the fracture toughness, K_{Ic}, follows from the load at which crack extension initiates. For the brittle material under consideration this will be the maximum load, since neither plasticity nor stable crack extension is expected to occur.

Experimental aspects

An important aspect of impact testing is the high rate of loading. Especially for brittle materials it is known that impact tests can be troublesome because of the short time to failure. At impact the striking tup will inevitably start to oscillate. In a material exhibiting plasticity, these oscillations will be diminished before the maximum load is reached, because there is more time available and also because energy is dissipated by plastic deformation in the specimen.

Tup stiffness

In the development of the experimental method tups were used with different load capacities, i.e. 220, 45 and 15 kN. Figure 3 shows examples of load-displacement curves for these three tups using the same impact

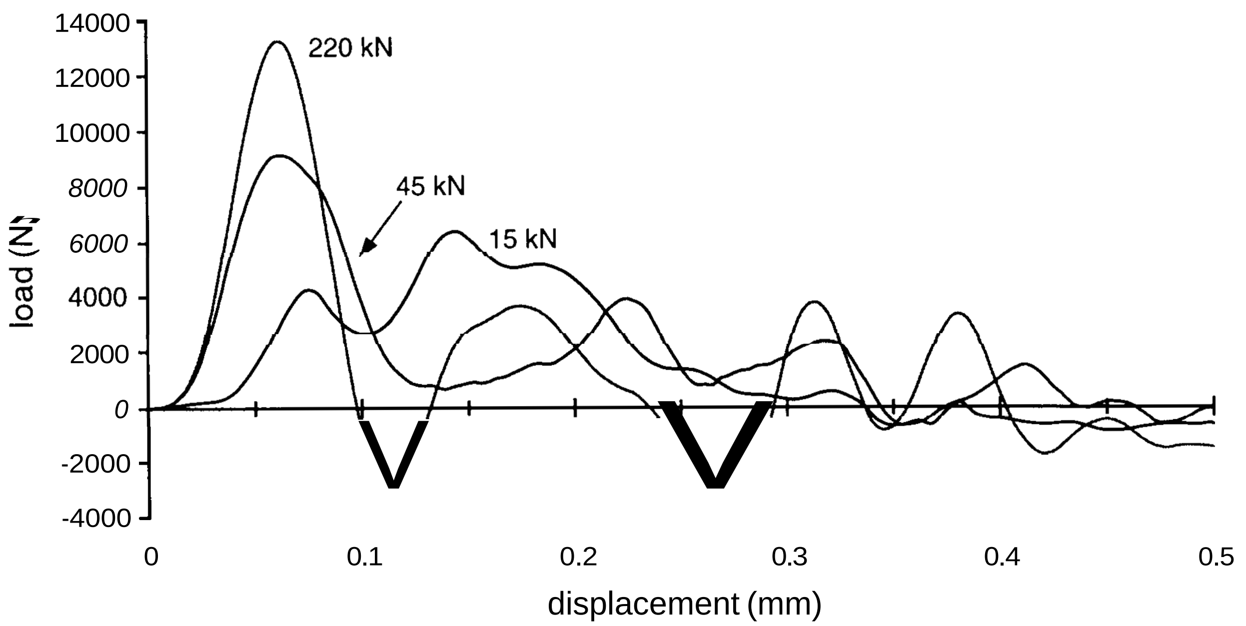

Figure 3: Load-displacement curves for the white cast iron at an impact velocity of 3.4 m/s using tups with different loading capacities

velocity of 3.4 m/s. Clearly, there is a large difference between the three curves. This can be related to the stiffness of the tups used. For a stiff tup the load rises quickly up to a high value, after which the load continues to oscillate for a considerable time. On the other hand in a compliant tup the oscillations have a much smaller amplitude and are almost damped before the maximum load is reached. It is noteworthy that in the oscillation occurring after specimen failure in the 220 kN tup a component is visible with a period ($\approx$ 35 μs) which more or less corresponds to twice the length of the tup ($\approx$ 110 mm) divided by the longitudinal wave velocity in steel, i.e. the material of which the tup is made (5900 m/s).

It is found that the stiffness of the tup also has an effect on the measured energy values. In Table 2 the energies are summarised for the three tests at 3.4 m/s shown in Figure 3. The 15 kN tup leads to a significantly higher energy. An explanation for this effect is probably related to the kinetic energy of the broken specimen after the impact. For example, an impact specimen that would attain a velocity of 3.4 m/s has a kinetic energy of 0.25 J, a value that cannot be neglected relative to the energy needed to fracture the specimen. For a low tup stiffness it is to be expected that a larger amount of kinetic energy is transferred to the specimen. The reason is that more elastic energy is stored in a compliant tup at the same load level.

TABLE 2

ENERGIES OBTAINED AT 3.4 M/S WITH TUPS HAVING DIFFERENT CAPACITIES

Tup capacity [kN]	Energy [J]
15	0.84
45	0.65
220	0.61

Impact velocity

The impact velocity also affects the test results, as can be seen in Figure 4 in which results are shown for 2 and 3.4 m/s, measured with a 15 kN tup. The oscillations are much more severe at the higher velocity, making realistic measurements of the maximum load difficult. Also the measured energy increases with velocity: 0.54 J at **2** m/s, 0.84 J at 3.4 m/s. The most likely explanation is the increase in kinetic energy of the broken specimen.

Reproducibility

It is noteworthy that the reproducibility of the tests seems to be relatively high considering the resemblance between two load-displacement curves for 2 m/s shown in Figure 4. This in spite of the presence of a notch

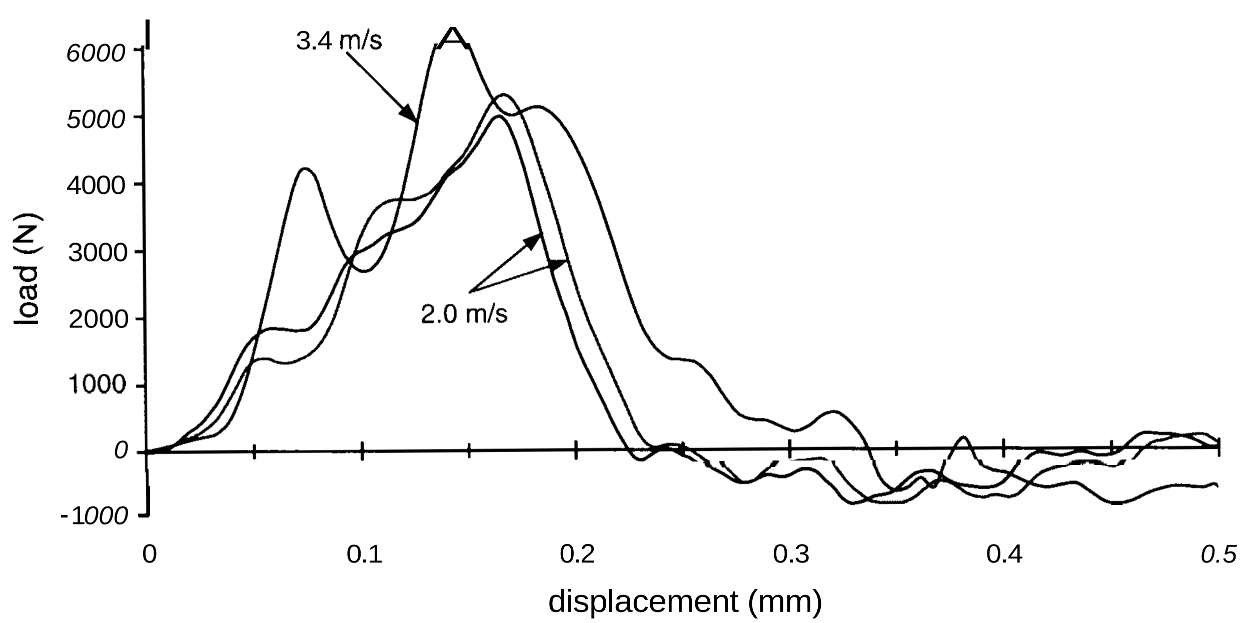

Figure 4: Test results for impact velocities of 2 and 3.4 m/s using a 15 kN tup

instead of a crack and the fact that the cast material, with its relatively coarse structure and possible porosity, may inherently show spread in behaviour.

Discussion

If the load at specimen failure needs to be determined, oscillations should be damped before this load is reached and thus a compliant tup is preferred. On the other hand, the measured energy will be too high due to a larger kinetic energy transferred to the specimen. In principle it is possible to correct for this effect if oscillations are damped before maximum load and if it may be assumed that the amount of energy stored in the tup at this load is completely transferred to the broken specimen. In case of a stiff tup the kinetic energy transferred to the specimen is lower, but is harder to predict due to the severe oscillations at the moment of failure.

Clearly, measuring at the lowest velocities would be preferable from an experimental point of view. However, this is only possible if the mechanical behaviour of the cast iron does not depend on the loading rate. At the moment there are still insufficient results to draw a definitive conclusion about this aspect.

EFFECT OF CARBIDE MORPHOLOGY ON TOUGHNESS

All tests on the effect of carbide morphology are performed using a 15 kN tup at 2 m/s. Fracture energy is calculated using the total measured work divided by the net section area of the specimen. Since the measurements only have a comparative nature, no attempt is made as yet to correct the results for the kinetic energy transferred to the specimen.

Macrostructure

Tests are performed on specimens (in the as-cast condition) having an equi-axed or columnar structure. The former were taken from the centre of the slab. The latter come from the outer slab region with the notch as close as possible to the cast surface with the notch normal to it. Table 3 gives an overview of the measured fracture energies and $K_{,,}$ values.

TABLE 3
RESULTS FOR THE COLUMNAR AND THE EQUI-AXED ZONE (I5 KN TUP, 2 M/S)

Structure	Fracture energy [kJ/m^2]	$K_{,,}$ [MPa$\sqrt{}$m]
Equi-axed	7.4; 6.1	70; 65
Columnar	5.6; *5.8*	49; *56*

On the fracture surface of the columnar specimens the orientation of the columns could be clearly observed. This indicates that the fracture path is affected by the presence of the carbide structures.

Hot Rolling

These experiments are performed on the same type of cast iron as the previous experiments, but the material comes from a different heat. Three slabs are involved which will be designated as I, II and 111. The slabs were all heated to 1050 "C and subsequently given the following treatment:

I. none (used as a reference)
II. rolling: 13.0 mm $\Rightarrow$ 12.6 mm
III. rolling: 12.5 mm $\Rightarrow$ 12.l mm; re-heating to 1050 "C; rolling 12.1 mm $\Rightarrow$ 11.0 mm

After this treatment the three slabs were cooled in air until 400 "C, kept at this temperature in a furnace for 90 minutes and finally cooled slowly in the furnace until room temperature. The latter treatment was included to optimise the structure of the matrix with respect to toughness. It should be mentioned that after the whole process slab II contained some cracks, while slab III contained several cracks.

Initially the intention was to apply reductions of 10% and 20% to slabs II and III respectively. However, since the rolling mill could not be accurately set to yield a pre-determined amount of deformation, this turned out differently. The total reductions of slabs II and III are estimated to be 3% and 12% respectively. Note that the intermediate thickness mentioned above for slab III is very approximate.

Rolling a slab will introduce rolling texture, since it will become longer in the rolling direction, while the thickness decreases. Therefore, it seems reasonable to expect that toughness values depend on the loading direction. For this reason impact specimens were taken from the slab with different orientations relative to the rolling direction. Figure 5 gives an overview. The specimen dimensions slightly varied, i.e. the cross section was only approximately **8** x 10 mm and the notch depth was between 1.3 and 1.9 mm. In all cases the loading was applied in the 8 mm direction.

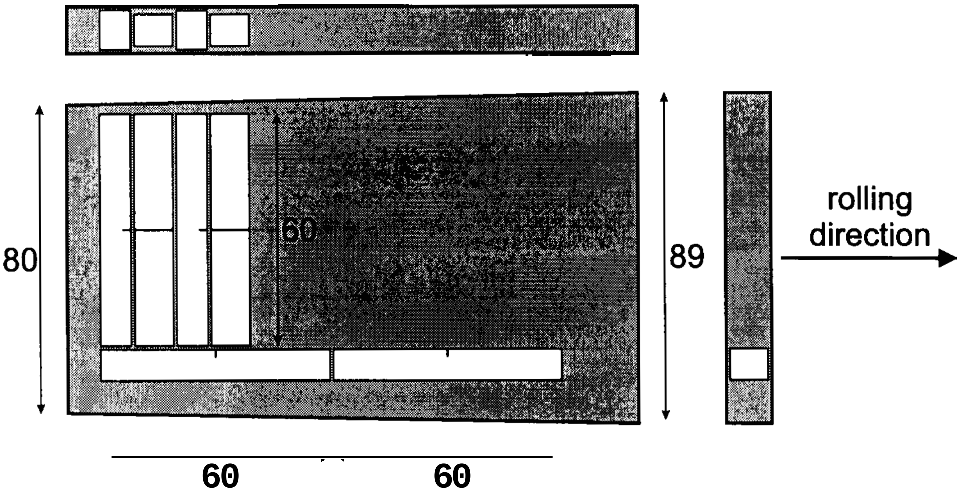

Figure 5: Specimen lay-out in each (rolled) slab

All the load-displacement curves obtained have a more rugged appearance than those of the tests described previously. This could suggest that crack growth is not continuous. As yet the cause is unclear: perhaps the material itself has a more coarse structure than that used for the other tests, or microcracks and/or porosity is present.

Table 4 gives an overview of the fracture energies and the *K*,, values determined. In the table it is indicated whether the fracture process is likely to be affected by cracks that were already formed during the rolling process. This could be recognised by the presence of parts on the fracture surface with a darker colour, resulting from the fact that this surface was formed at high temperature. Furthermore, in the table a best estimate for both fracture energy and *K*,, is given.

TABLE **4**

FRACTURE ENERGIES AND K_{Ic} VALUES FOR HOT ROLLED WHITE CAST IRON AT 2 M/S

Slab	Reduction		Loading direction: ⊥ to slab surface			// to rolling direction			⊥ to rolling direction		
I	0%	Energy [kJ/m^2]	6.8	8.2	**7.5**	8.4	7.6	**8.0**	7.8	8.8	**8.3**
		K_{Ic} [MPa$\sqrt{m}$]	79	84	**82**	70	86	**78**	77	95	**86**
II	3%	Energy [kJ/m^2]	10.0	10.1	**10.1**	7.4	9.4	**8.4**	9.4	11.8	**11.8**
		K_{Ic} [MPa$\sqrt{m}$]	84	94	**89**	70	88	**79**	38	75	**75**
III	12%	Energy [kJ/m^2]	9.7	10.3	**10**	9.7	9.8	**9.8**	11.5	12.0	**12.0**
		K_{Ic} [MPa$\sqrt{m}$]	75	83	**79**	78	82	**80**	40	81	**81**

: few cracks   : many cracks   12.3: test data   **12.3**: best estimates

Discussion

The results from Table **3** clearly indicate that both the fracture energy as well as the ***K,,*** is higher for the equi-axed than for the columnar structure. A possible explanation for this is that in the columnar specimens the chromium carbides are orientated parallel to the notch and thus at the notch tip the highest tensile stress is normal to the carbides.

The results for the rolling experiments (Table **4)** suggest an increase of the fracture energy with reduction percentage for all loading directions. This tendency, however, is not reflected in the ***K,,*** results. Possibly, the carbide network is indeed broken-up by rolling the cast iron. As a result the propagation of cracks would be hindered, leading to a higher fracture energy. On the other hand, ***K,,*** is a measure for the load needed for crack initiation. This load level would be reached if at any point in the vicinity of the notch tip the partially broken-up carbide structure is loaded critically. Furthermore, since cracks have been observed macroscopically, they are also likely to be present on a micro scale. Such cracks would counteract a possible increase in fracture toughness caused by the rolling process.

The results do not suggest a clear correlation with the loading direction. Perhaps this is due to the fact that for these experiments the presence of an equi-axed or columnar structure in the different specimens was not taken into account. Finally it should be noted that the changes in microstructure caused by rolling the material have not yet been evaluated.

CONCLUSIONS

1. Impact tests on a brittle material such as high-chromium white cast iron are strongly affected by the stiffness of the tup. A relatively low stiffness is advantageous for determining fracture energy and load.
2. The energy found through impact tests of the cast iron increases with velocity. This is attributed to an increase in kinetic energy transferred to the broken specimen.
3. Impact results for the cast iron reproduce well in spite of the fact that a notch is used instead of a crack.
4. Both the fracture energy and *K,,* are found to be higher in cast iron having an equi-axed structure than in material with a columnar structure.
5. Hot rolling the white cast iron seems to increase the fracture energy somewhat. No consistent effect on ***K,,*** is found.

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