The influence of carbide morphology on the upper-shelf fracture toughness (J_{Ic}) of carbon steels

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Abstract

UNI C 15, C 45, and C 85 plain carbon steel specimens were prepared for J-integral testing in the furnace-annealed, normalized, and spheroidized conditions at an upper-shelf temperature of 225 °C. Furnace-annealed and normalized steels behave similarly in regards to toughness, whereas spheroidized steels exhibit higher levels of toughness and ductility. Since fracture in low and medium C lamellar pearlite steels occurred predominantly in the ferrite, fracture toughnesses were associated with the distribution of pearlite colonies within the ferrite and the resulting contained plastic deformation, which grows as the average free ferrite parth decreases. Also J_{Ic} values in spheroidized structures resulted depended mainly on the mean free path between carbides with a decrease in spacing resulting in lower toughness values.

Riassunto

Influenza della morfologia dei carburi sulla tenacità alla frattura duttile (J_{Ic}) degli acciai al carbonio

Sono state effettuate determinazioni dell'integrale-J applicato durante la propagazione di frattura in campioni di acciaio tipo UNI C 15, C 45 e C 85. Le prove, condotte su provini trattati termicamente mediante ricottura completa, o normalizzazione, o ricottura di sferoidizzazione, sono state effettuate a 225°C per garantire per tutti i tipi di acciaio e di trattamento, il raggiungimento del pianerottolo superiore della curva resilienza-temperatura.

Gli acciai sottoposti a normalizzazione o ricottura completa presentano caratteristiche di tenacità a frattura molto simili; la tenacità e la duttilità risultano notevolmente superiori quando la cementite è presente sotto forma di sferoidi.

In considerazione del fatto che, nel caso di acciai a basso e medio tenore di carbonio con struttura lamellare della perlite, la frattura si è propagata quasi esclusivamente attraverso le zone ferritiche, i valori di tenacità (J_{Ic}) sono stati messi in correlazione con i parametri di distribuzione delle colonie perlitiche all'interno della matrice e interpretati ipotizzando che, al diminuire del cammino libero medio nella ferrite, si accentui un fenomeno di deformazione plastica contenuta che porta ad un decremento della tenacità. Anche i valore di J_{Ic} per gli acciai allo stato sferoidizzato dipendono in massima parte dal cammino libero medio tra i carburi. Tale relazione di dipendenza ci ha permesso di riscontrare che, a minori distanze fra i carburi, corrispondono valori inferiori della tenacità a frattura.

Commonly used plain carbon steels are characterized by a ferrite-pearlite microstructure. The amount of each of these phases depends chiefly on the carbon content and cooling rate from the austenitizing temperature. Increases in either of these factors increase the amount of pearlite.

Much attention has been focussed upon effects that varying ratios of ferrite to pearlite have on mechanical properties. For example, in low carbon steels containing up to 0.25% C, it has been shown that increasing the carbon content gives a marked increase in the ultimate tensile strenght (UTS) without the yield strength (YS) being comparably affected [1-3]. This indicates that ferrite controls the YS of these low carbon materials and the pearlite controls the UTS. With medium carbon steels of about 0.4% C, the ferrite also plays the same important role in relation to the mechanical properties. When the ferrite grain size decreases, or it is solid solution strengthened by elements such as Mn and Si, the YS goes up. In higher carbon steels near the eutectoid composition, an increase in the amount of pearlite yields comparable increases in both the UTS and YS. Here, the controlling factor is the ferrite-cementite interlamellar spacing in the pearlite. A coarser pearlite leads to decreases in both the UTS[4,5] and the strain needed for fracture[6,7]. A thinner cementite lamella is more resistant to cracking. With thinner cementite, the corresponding decrease in ferrite thickness (carbide spacing) reduces stress concentrations at slip intersections within the ferrite-carbide interface, making decohesion or carbide cracking less likely. Thus the ductility of fine pearlite is much greater[5]. It has also been noted that as the pearlite content increases the ductility decreases.

With regard to the mechanical properties of these steels in the presence of a triaxial stress typical of a notch or a fine crack in a thick enough component, it is known that an increase in the amount of perlite leads to a corresponding increase in the ductile to brittle transition temperature (DBTT) and a decrease in the amount of absorbed impact energy[2,4,6,8]. In particular, in medium carbon steels an important role is again played by the ferritic grain size and also by the pearlite interlamellar spacing, since decreases in their size lower the DBTT. Much less information is presented in the literature with regard to this class of steels in the spheroidized condition. Pickering[5] states that when spheroidized these steels acquire a better ductility than in the normalized state in the whole range of carbon contents up to 0.8% C. There is a corresponding increase in notch toughness and a decrease in the DBTT[9].

Research reported in the literature about the fracture resistance of carbon steels in the presence of sharp cracks has been mainly focussed on their behaviour at temperatures below the DBTT. Instead, little work has been performed at temperatures at which ductile fracture predominates, except in the case of low C steels (see ref. 10 for a review of data). In fact, carbon steels reach such a high toughness that this property cannot be characterized in a practical manner by a linear elastic fracture mechanics approach. It would entail testing specimens of considerable thickness to provide the constraint necessary for the plane strain conditions required for linear elastic fracture mechanics. This becomes especially true at the upper-shelf test temperature where the Charpy-V notch impact energy is at a maximum and the fracture appearance is 100% fibrous. Nor have these steels been characterized with varying carbon contents in the normalized versus furnace-annealed versus spheroidized contitions by a

TABLE I - Chemical composition of the steels

Steel	%C	%Mn	%P	%S	%Si	%Cr	%Ni
C 15	0.14	0.93	0.009	0.011	0.09	—	traces
C 45	0.42	0.83	0.023	0.010	0.17	0.03	traces
C 85	0.86	0.51	0.033	0.015	0.25	0.02	traces

J-integral approach, which allows the use of smaller testpieces. Only Curry and Pratt[11] have recently published ductile fracture COD and J_{Ic} results obtained with two 0.29% C, low-alloy steels with different MnS content, tested in the full-annealed, normalized, or partially speroidize-annealed conditions. It is the purpose of this investigation to offer a first complete attempt for such a characterization.

Three types of carbon steels were selected, namely AISI 1018, 1045, and 1090, which are roughly equivalent to UNI C 15, C 45, and C 85 steels (the latter codes have been used for identification throughout the paper, since they are more similar to other European designations). All the steels have been tested in the funace annealed and spheroidized conditions. Normalizing was performed only on the C 15 and C 85 steels for which the resulting microstructures are less dependent on either prior heat-treating and mechanical working history or sample dimensions.

Experimental part

The steels, produced in the U.S.A., were received as 25.4 × 50.8 mm rectangular section bars. Their chemical composition is reported in Table I.

Three-point bending specimens 254 mm long were cut from the bars and heat-treated as reported in ref. 12 to obtain the desired microstructures. Final machining, as well as notching and fatigue precracking in the LT direction, were performed after heat- treating. Details of the procedure are also described in ref. 12.

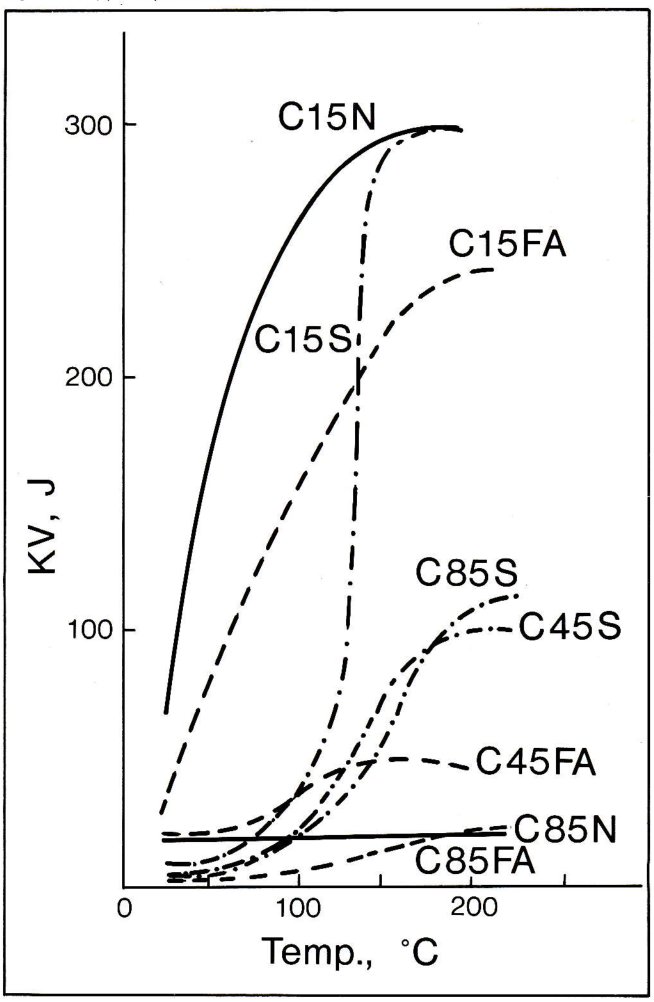
The multiple specimen heat tinting technique, as described in ASTM Standard E 813-81, was adopted to determine the critical applied J-integral value at the onset of crack advance (J_{Ic}). Standard tensile and J-integral tests were conducted at 225 °C by the use of an asbestos furnace about the loading apparatus. In fact Charpy-V impact transition curves (Fig. 1) had shown that, at such a temperature, the upper shelf region is reached and the fracture appearance is 100% fibrous for all the steels. Each specimen was loaded in a 250 KN closed loop servo-hydraulic Instron machine in a fixture with a span of 203 mm. Different deflections were obtained using a cross-head rate of 0.05 mm/s; unloading was performed at a faster rate to eliminate the possibility of creep occurring.

The use of clip-gages for monitoring bend specimen deflections was prevented by the furnace set-up. Therefore J-integral determinations were performed by summing up the plastic and elastic components of J (J_{pl} and J_{el}). The former was calculated by the formula[13], $J_{pl} = 2A_{pl}/Bb$, where B is the thickness and b the remaining ligament of the precracked specimens. A_{pl} was taken as the plastic portion of the area under the load-cross head displacement record. J_{el} was calculated as K^2/E, with K being the stress intensity factor computed according to the ASTM specification E 399-81[°]

Corresponding values of J and crack extension (Δa) yielded the J resistance curve from which the J_{Ic} data was obtained at its intersection with the blunting line of equation, $J = 2\sigma_Y \Delta a$. σ_Y was taken as the average of UTS and YS at 225 °C.

Polished cross sections of the samples were examined by quantitative metallograply using optical techniques[12] to determine the volume fraction, particle size, mean free path, and interparticle spacing of cementite spheroids in spheroidized steels, as well as the average free ferrite path between pearlite colonies in C 15 and C 45 steels in the furnace-annealed and normalized conditions.

Fig. 1 - Charpy-V impact transition curves



[°] It was also verified that the use of the formula, $J = 2A/Bb$, with A being the total area under the load-cross head displacement records, yielded J values less than 1% different from those derived as above illustrated.

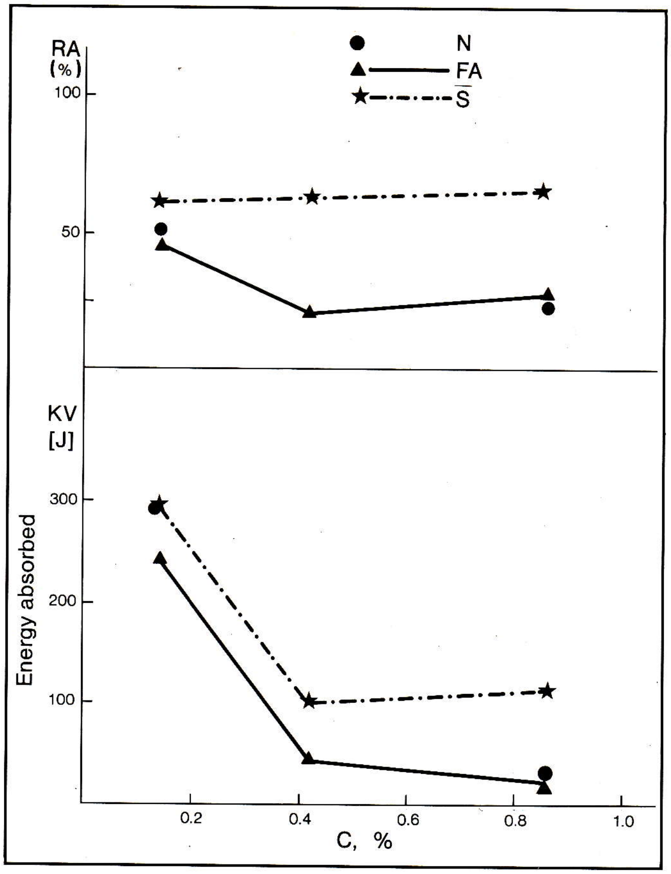


Fig. 2 - Reduction of area (a) and impact energies (b) vs. C content.

Results and discussion

The J_{Ic} data at 225 °C are listed in Table II along with tensile and impact test results obtained at the same temperature.

Various correlations among mechanical test results, carbon contents, and microstructural parameters were tried to clarify the influence of composition and heat-treatment on the characteristics of the steels. Some of the resulting plots are presented hereafter.

Reductions of area (RA) in tensile tests and absorbed energies in Charpy impact experiments are reported in Fig. 2 as a function of the carbon percentages and heat-treatments. The beneficial effect of spheroidizing upon both these properties is evident. It can also be seen that the variation in both the above characteristics indicates a similar influence of increasing carbide volume fractions in similarly heat-treated steels; i.e. a marked decrease passing from the C 15 to the C 45 steels and then no change up to the eutectoid composition. A conspicuous exception to this behaviour is represented by the constancy of the RA in speroidized steels upon increasing the carbon content. If such a property is to be considered an accurate indication of the ductility of an alloy, the above findings are in marked contrast with results reported by Pickering[5] and Liu and Gurland[14], who instead observed a decrease in ductility upon increasing the carbide volume fraction in globular cementite steels.

Their experiments however, were performed at room temperature; this temperature did not correspond to the upper-shelf, at least for the steels with either a medium or a high carbide content.

On the other hand, the data reported in Table II demonstrate that the reduction of area is a tensile property which shows an almost linear correlation with the fracture toughness, if results pertaining to each of these steels are considered separately.

The variation in J_{Ic} values with increasing carbon contents is plotted in Fig. 3. The spheroidizing treatment increased the toughness of all the steels and

TABLE II - Tensile, impact (KV) and J_{Ic} data at 225 °C.

Steel	J_{Ic} (MJ/m^2)	KV (J)	YS (MPa)	UTS (MPa)	RA (%)
C 15 N	0.207	297	204	521	50
C 15 FA	0.197	244	160	480	45
C 15 S	0.286	297	149	437	60
C 45 FA	0.055	43	285	733	20
C 45 S	0.097	99	291	496	64
C 85 N	0.046	22	498	962	24
C 85 FA	0.059	20	233	654	28
C 85 S	0.163	114	198	481	65

N = normalized; FA = furnace-annealed; S = spheroidized it is clearly indicated here. Such a finding further demonstrates that the J-integral approach is particularly sensitive in distinguishing the difference between different heat-treatments.

It is to be also noted that J_{Ic} results for C 15 and C 85 steels in the normalized and furnace-annealed conditions indicate a negligible influence of the pearlite interlamellar spacing upon the fracture toughness of the steels at the upper-shelf level, as was also confirmed by ductility and Charpy impact data.

Scanning electron microscope photographs of the fracture surfaces can be seen in Figures 4, 5, and 6 alongside transverse section optical micrographs. In the normalized or furnace annealed steels extensive cracking across the lamellar pearlite colonies is evident only in the eutectoid steel, whereas both the C 15 and the C 45 steels fractured with predominant crack propagation across the ferrite. This is particularly interesting for the medium carbon steel since it indicates a case of contained ductile fracture(15) in which plastic flow is localized in the softer constituent (ferrite) to alleviate high elastic stress concentrations due to the harder constituent (lamellar pearlite). Such a strain concentration can cause fracture to occur at a lower level of macroscopic strain than if the strain were homogeneously distributed. Thus fracture occurs earlier in the deformation process(16).

This is clearly evidenced in Fig. 7 in which fracture morphologies of C 85 and C 45 steels in various heat-treatment conditions are compared at a higher magnification. The furnace-annealed steel rupture surfaces show that fracture across the pearlite colonies intervened in the C 45 steel at a lower level of local deformation in respect to that reached by the corresponding C 85 steel, as further demonstrated by the shallowness of the dimples appearing in the ferrite constituent of the former one.

The above considerations coupled with the difference in microstructure can be applied to explain the large drop in J_{Ic} values indicated in Fig. 3 upon passing from the C 15 furnace annealed steel to the C 45 one. In fact, the average free ferrite path between pearlite colonies is 87 μm for the former steels as compared to 61 μm for the latter one. Furthermore, in the C 45 steel, zones in which the thickness of the ferrite network decreases to 10 μm ca. are frequent.
The above reported fractographic analysis also explains why there is no practical difference in fracture thoughness between the furnace-annealed C 45 and C 85 steels.

The same rationale can be used to explain the variation of J_{Ic} data in the case of steels in the spheroidized condition, as well as the higher fracture toughness of spheroidized pearlite-ferrite structures in respect to the

Fig. 3 - J_{Ic} values versus carbon content.

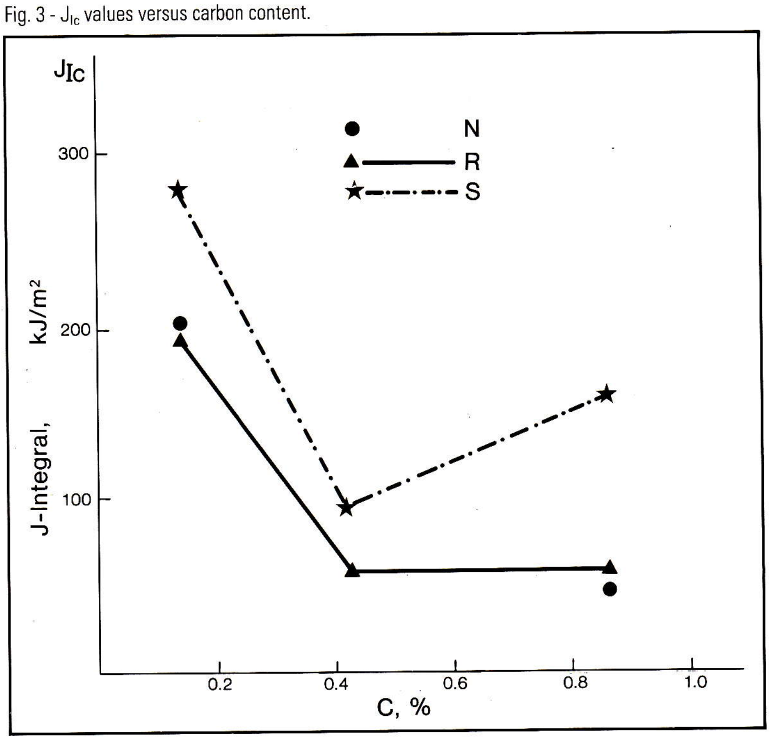


Fig. 4 - Microstructures and fractographs of C 15 steel in the Normalized (N), Furnace Annealed (FA), and Spheroidized (S) conditions.

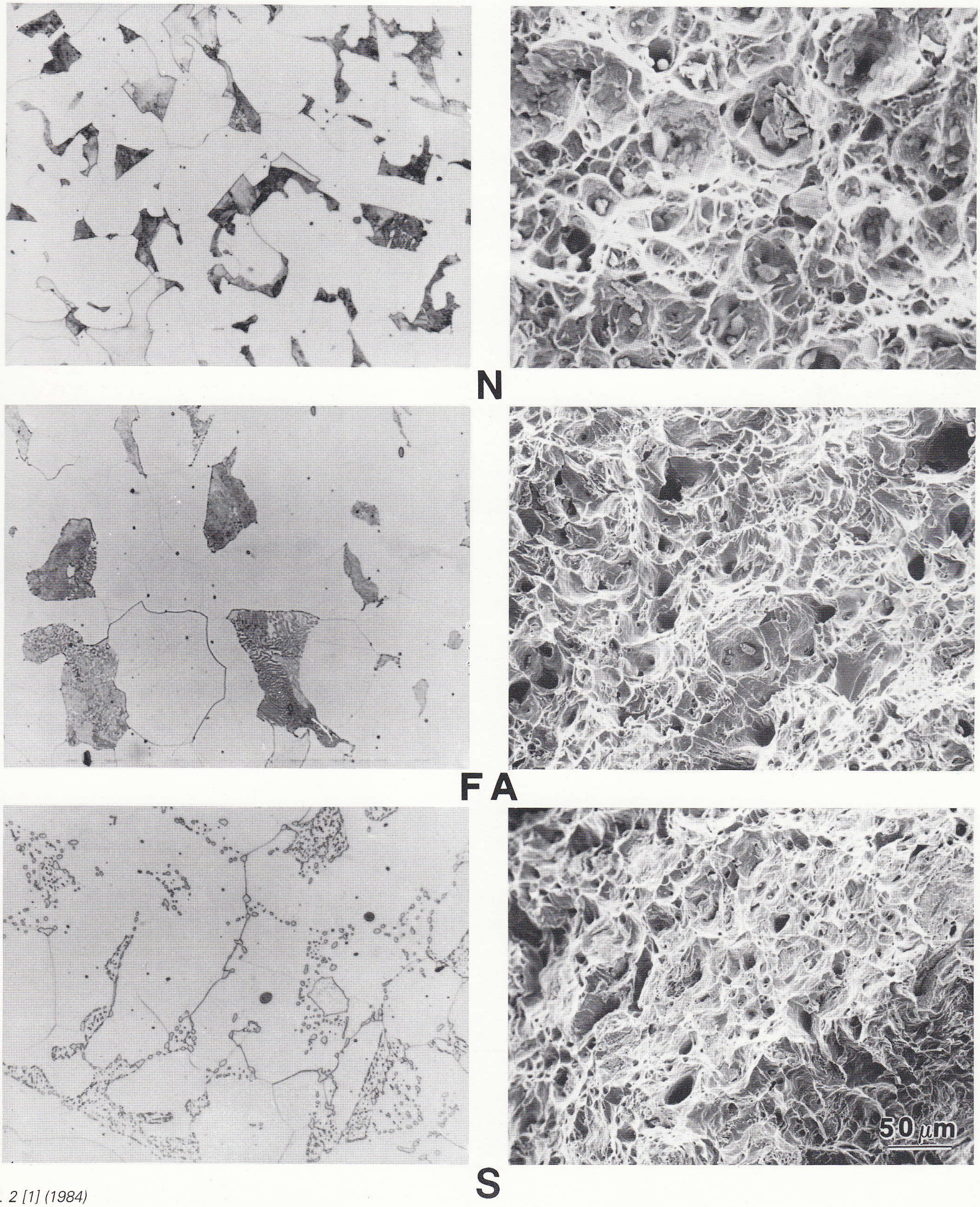


Fig. 5 - Microstructures and fractographs of C 45 steel in the Furnace Annealed (FA) and Spheroidized (S) conditions.

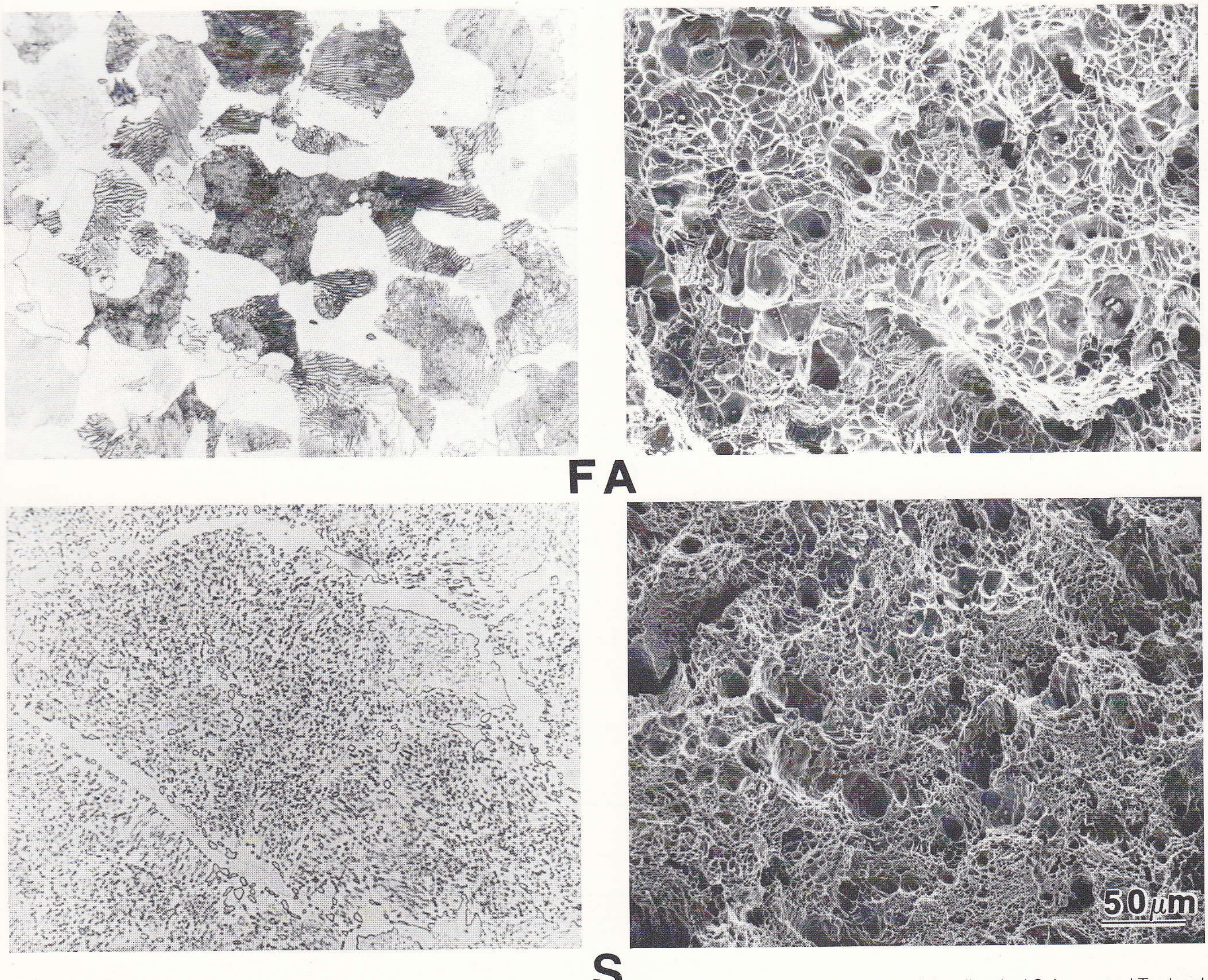


Fig. 6 - Microstructures and fractographs of C 85 steel in the Normalized (N), Furnace Annealed (FA), and Spheroidized (S) conditions.

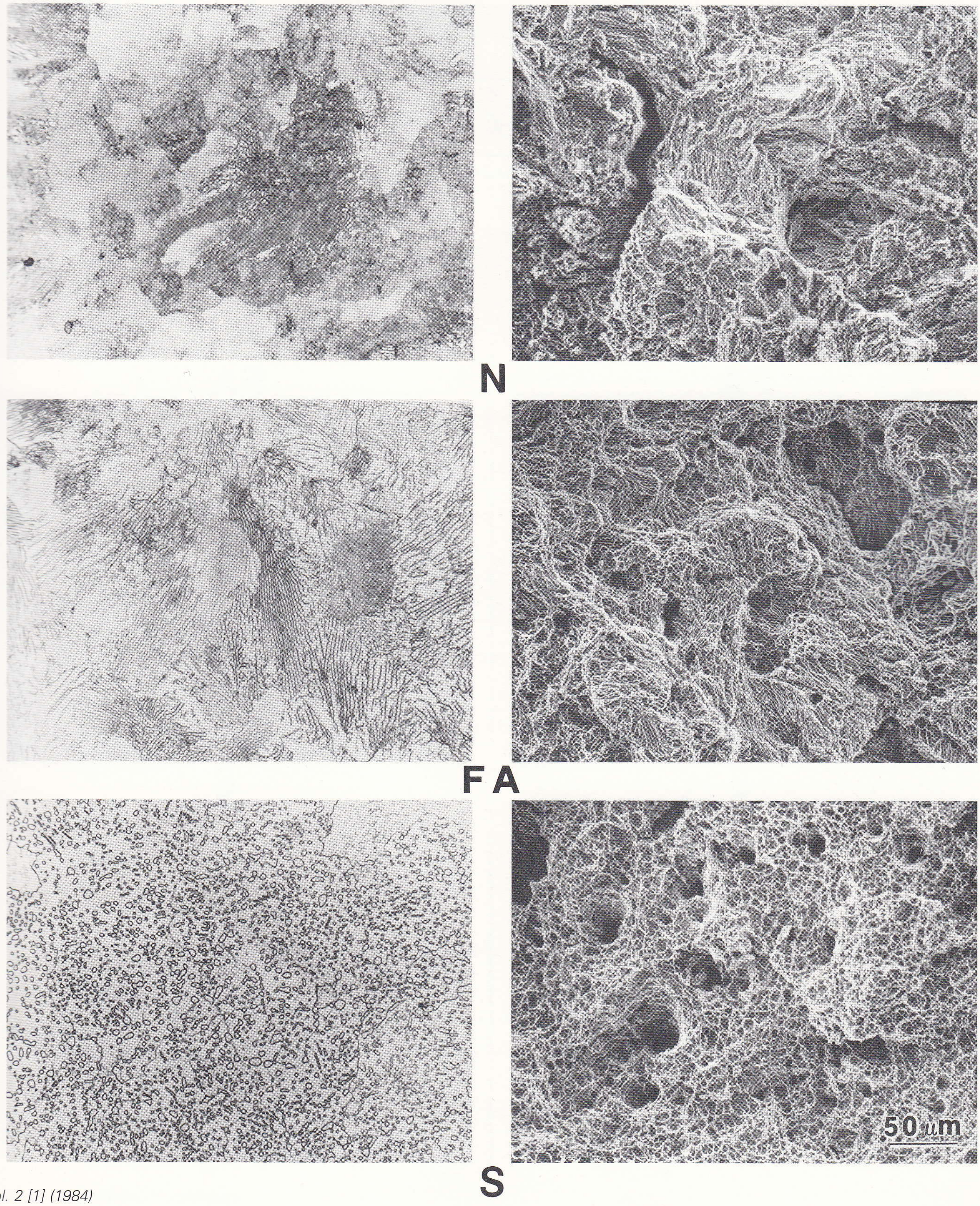
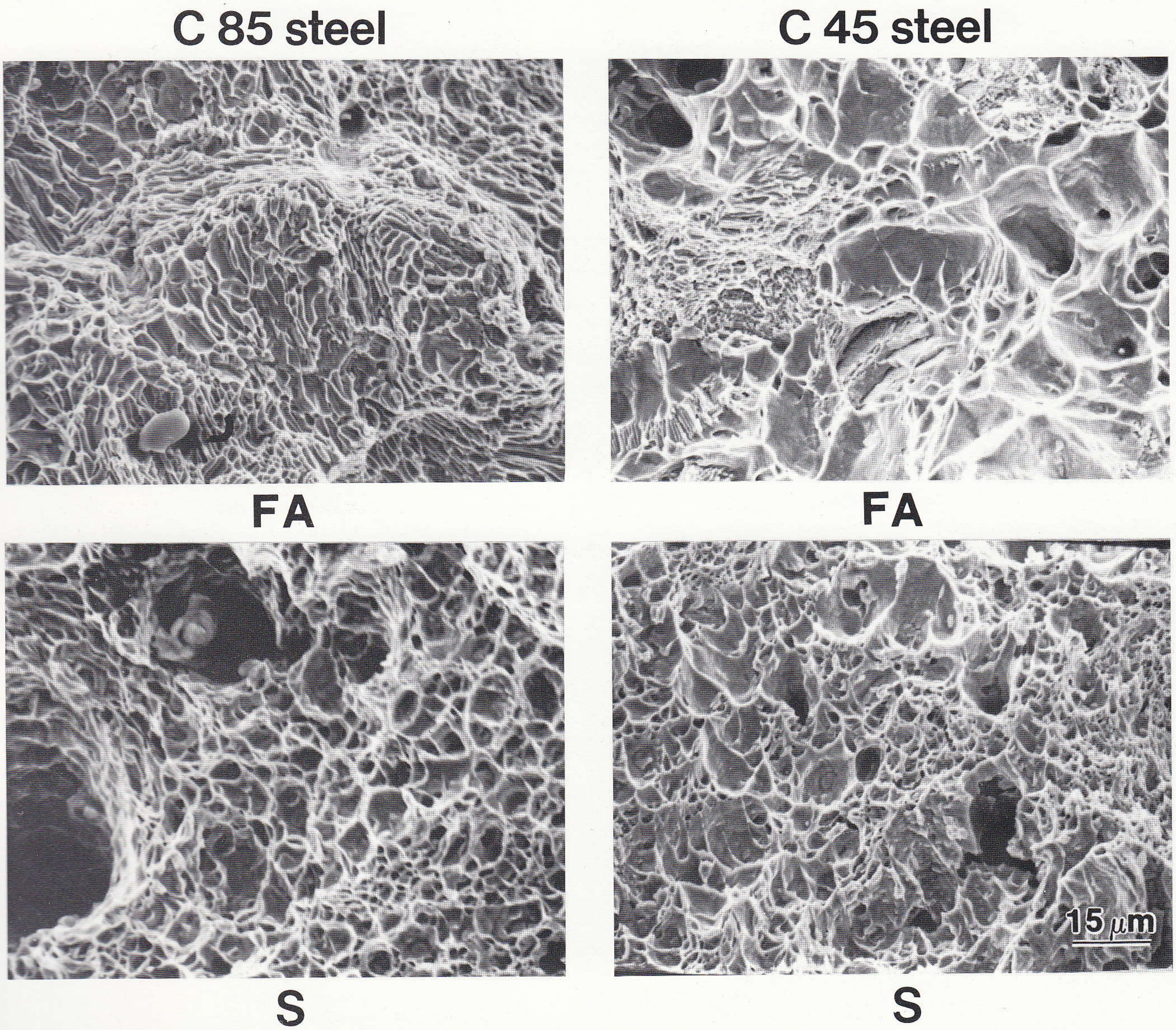


Fig. 7 - Comparison between fracture morphologies of C 85 and C 45 steels in different heat-treatment conditions.



lamellar pealite ones. In fact, the spacing between the Fe_3 C particles restricts the amount of deformation in the plastic zone about the crack tip, as demonstrated by Fig. 8 which indicates a direct correspondance between fracture toughness and mean free path between carbides, the C 45 steel being characterized by the lowest values of both the above quantities. Furthermore, it is known that a steel will undergo a large amount of deformation prior to the nucleation of voids (to be later linked to the crack tip), because carbides do not crack or decohere at low strains.

However the strain required for spheroidal carbide cracking is higher than that necessary for cracking pearlitic lamellar carbides[5]. Therefore, considerable plastic deformation occurs in a spheroidized steel before material separation intervenes. The decohesion of the Fe_3C particles eventually takes place creating microvoids which coalesce. The equi-axed dimples present on the fracture surfaces of spheroidized C 45 and C 85 steels (Fig. 7) are thus justified, since their formation is evidence that material separation occurred because of appreciable plastic deformation.

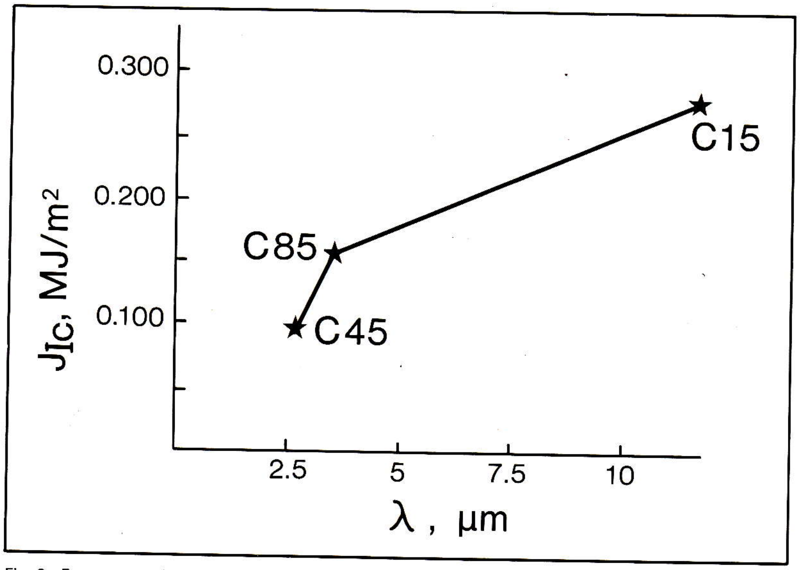


Fig. 8 - Fracture toughness vs the mean free path between carbides (λ) in spheroidized steels.

Conclusions

i) Plain carbon steels in the furnace-annealed and normalized conditions exhibit similar levels of upper-shelf fracture toughness and ductility in the 0.14 to 0.86% C range. These steels show greater fracture toughness and ductility levels in the spheroidized condition.

ii) The differences in the characteristics of the pearlite in the furnace-annealed versus normalized condition have little effect on fracture toughness and ductility.

iii) As the mean free path between spheroidal carbides or the average free ferrite path between pearlite colonies decrease, the fracture toughness decreases. Particle size has comparativelg little influence on fracture toughness.

iv) The presence of carbides or pearlite islands is believed to highly conconcentrate the plastic deformation in the ferrite causing failure at lower toughness levels. These levels are related to the spacing between these constituents (see item iii).

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