JOURNAL OF MATERIALS SCIENCE 18 (1983) 2181–2199

The effect of microstructure and stress state on the fracture behaviour of wood

S. W. J. BOATRIGHT
Aspen Graphics, 124 Tabernacle Street, London, EC2, UK

G. G. GARRETT
Department of Metallurgy, University of the Witwatersrand, Johannesburg, Republic of South Africa

Based on a general fracture mechanics approach, this paper attempts to correlate macroscopic fracture properties with micromechanical behaviour in wood. In the first part, the influence of orientation, cell size and other structural features of wood on toughness, fatigue resistance and fracture morphology is described. Subsequently, experimental observations of crack tunnelling effects are reported which, coupled with direct crack tip strain measurements and a verification of the influence of specimen thickness on fracture properties, confirm the existence of stress state variations across a crack front in wood, and the corresponding effects on fracture behaviour. Irreversible crack-tip deformation modes are identified, involving intercellular debonding, cell twisting and buckling, and the role of stress triaxiality is discussed. A strain criteria for the fracture of pre-notched bulk wood specimens is developed, based on an observed linear relationship between notch root radius and crack opening displacement for fracture initiation.

1. Introduction

For its intended purpose, wood, as a natural composite material, is probably superior to anything attempted by advanced materials science. How many self-confessed composites specialists could, for example, successfully take on the job of designing and manufacturing a tree? Although strong similarities exist between nature's own composite and artificial fibre-reinforced polymers, the microstructural design of wood is unique and exceedingly complex compared with man-made composites.

Even with the advent of sophisticated materials technology, wood remains a very popular constructional material: annually the world usage is approximately 10^9 tons, and more than twice that of steel. Therefore, it is somewhat surprising, and perhaps a consequence of its complexity that, relative to metals, plastics and composites, little effort has been made by materials scientists on research into the mechanical properties of wood. This is particularly true in the field of fracture, which has relevance not only to the structural use of timber, but also to the important processes of machining and cutting.

One reason for this relative lack of knowledge concerning the fracture behaviour of wood has been that wood has such good specific toughness properties that the designer has had little recourse to rigorous brittle fracture criteria. In addition, attempts to relate gross mechanical properties to the micromechanical behaviour have been hampered by insufficient knowledge of the microstructure itself. However, in recent years much of the nature of this complex, cellular composite has been established and success in relating the structure and mechanical properties of wood is now more readily achieved.

A further factor arises from the inhomogeneity of the bulk structure itself. At the present time, the forestry and wood processing industry certainly does not have the control over the quality of its product that, say, the steel or the plastic industries have over theirs. Structural timber is

0022–2461/83 $03.00 + .12

liable to considerable variation in its mechanical properties, and worse still, will contain many potentially strength-reducing flaws. Checks, for example, are cracks, usually in the radial-longitudinal plane, and typically caused by differential shrinkage during seasoning. Knots, on the other hand, often involve considerable grain disturbance and also (if dead) a loss in the load-bearing area.

To make the most efficient use of timber of all qualities, a means of strength assessment is essential. Failure is generally associated with such irregularities in the timber structure, and so strength assessment relies both on the detection of flaws and a knowledge of their weakening effect. There is some attraction, therefore, in attempting to employ a quantitative approach in order to evaluate the strength of timber.

Wood, if considered in elements substantially larger than the cell and over relatively short loading periods, can be approximated to a linear elastic continuum [1] and, on this basis, Attack and co-workers [2] were apparently the first to attempt to apply linear elastic fracture mechanics (LEFM) to wood failure. They were primarily concerned with the energy of the pulping process, but showed that the product of the fracture stress and (crack length)$^{1/2}$ is approximately constant. Schniewind and Pozniak [3], for example, subsequently confirmed the independence of K_{Ic} from crack length, indicating that fracture toughness could well be useful as a material constant. A general fracture mechanics approach has, therefore, been used in the work reported here, primarily as a base for correlating macroscopic fracture properties with micromechanical behaviour.

The first part of this paper studies the role played by various microstructural features in determining the fracture behaviour. Fatigue results are presented as a convenient method for quantitatively controlling sub-critical crack growth. The effect of the stress state on the fracture behaviour of wood is then examined and finally, based on an experimental approach, a realistic criterion for fracture is developed. It is first necessary, however, to briefly introduce the material itself.

2. Structure and microstructure

Several well-written articles [1, 4–6] are available which review the literature to varying degrees, and which should be consulted for a more in-depth background to the current state-of-knowledge concerning the structure–property relationships in wood. In brief, and greatly simplified, wood consists of discontinuous, overlapping, hollow fibres (or tracheids) composed of cellulose, which vary in length in the range 1 to 4 mm, and have a length to diameter ratio of about 100:1. They are primarily aligned parallel to the axis of the tree trunk, and are cemented together within a continuous matrix (known as the middle lamella) by an amorphous, organic compound called lignin.

The wood fibres represent the reinforcing element and carry the main load, having an identical function to synthetic fibres in man-made composites. But there the straightfoward similarity ends, since the fibres themselves are complex, tubular structures with multi-layered, laminated walls telescoped onto, and bonded with, each other (Fig. 1). These cell wall layers (or laminae) are made up of either parallel or criss-crossed strands of cellulosic microfibrils, each about 20 nm in diameter and in excess of a micrometre in length, embedded in a matrix consisting of lignin plus hemicellulose plus non-crystalline cellulose. The cell wall is generally made up of the primary (P) layer, where the microfibrils are randomly arranged, together with several secondary layers (S_1, S_2 and S_3), Fig. 1a. When growth is more rapid, typically in spring, the cell walls of the so-called "early" wood are thinner with larger central cavities (or luminae), resulting in a lower density structure. From hereon these cells will be referred to as large (diameter) cells. Cells produced later in the year ("late" or summer wood) are of smaller diameter with thicker walls (hereon called small cells). The demarcation between the two types of cell defines the visible, annual growth rings. Other structural features having specialized growth functions include pits, which connect the fibre luminae, and ray cells (about 5% of the total fibres) which traverse the structure in a radial direction (Fig. 1b).

3. Experimental details

It was considered unnecessary to study all six possible crack propagation systems: each of the groups (LT, LR), (RL, RT) and (TL, TR) has the same crack plane and within these groups the toughness values are very similar [7]. In the work described here the RL, TL and LT systems, Fig. 2, were chosen as representative, and these, in fact, simulate the common defects, shakes, checks and knots, respectively. However, while shakes and checks more normally exhibit mode II crack opening, under conventional structural loading,

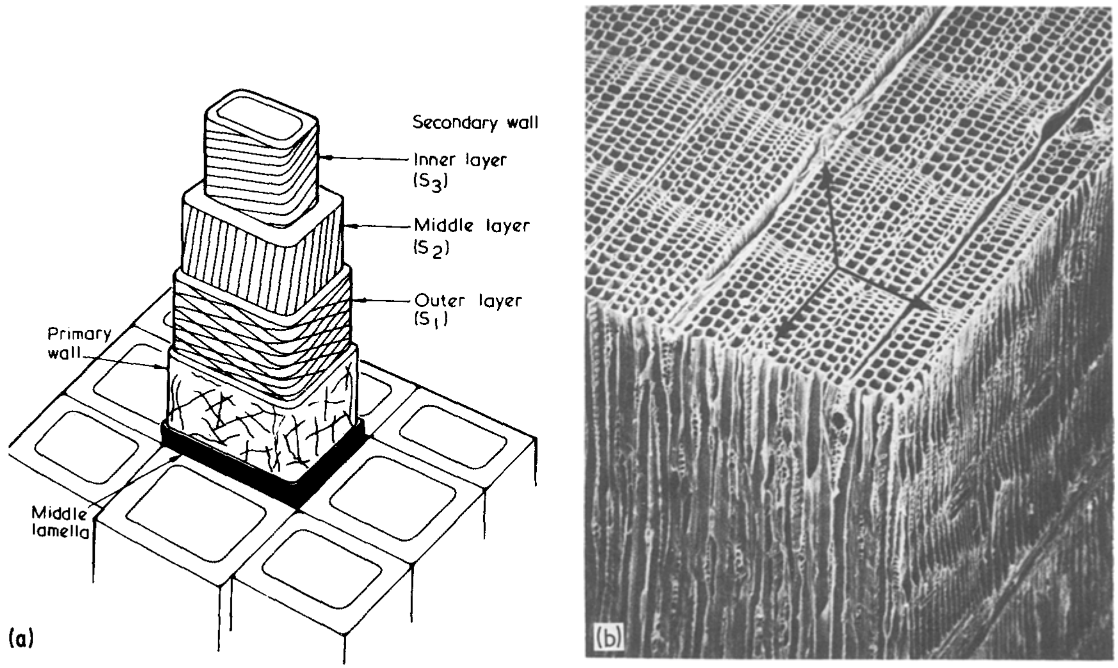


Figure 1 (a) Simplified, schematic illustration of the structure of a cell wall in wood, showing the orientation of the microfibrils in each layer. (After Dinwoodie, [4]). (b) Block of oak timber enlarged to show the types and distributions of cells on the three principal axes. (Courtesy of J. D. Barratt, from *Phil. Trans. Roy. Soc. Lond.* A299 (1981) 217).

studies reported here were confined to the much simpler mode I crack opening, in order to provide a suitable parameter quantifying the fracture behaviour of wood at a first level.

All tests were carried out using South African pine (*Pinus radiata*) and the material chosen for test specimens had even growth rings, was defect free, and had straight grain. Prior to testing all specimens had been conditioned at 20° C and 65% approximate relative humidity for a minimum of 28 days. The average moisture content measured was 12.2% based on oven-dry mass.

Fracture toughness tests were carried out on single edge notched (SEN) specimens, primarily in 4-point bending; some (thin specimen) tests were conducted in tension. Initial cracks were produced using a sharp blade (though some comparative experiments were carried out using fatigue cracks).

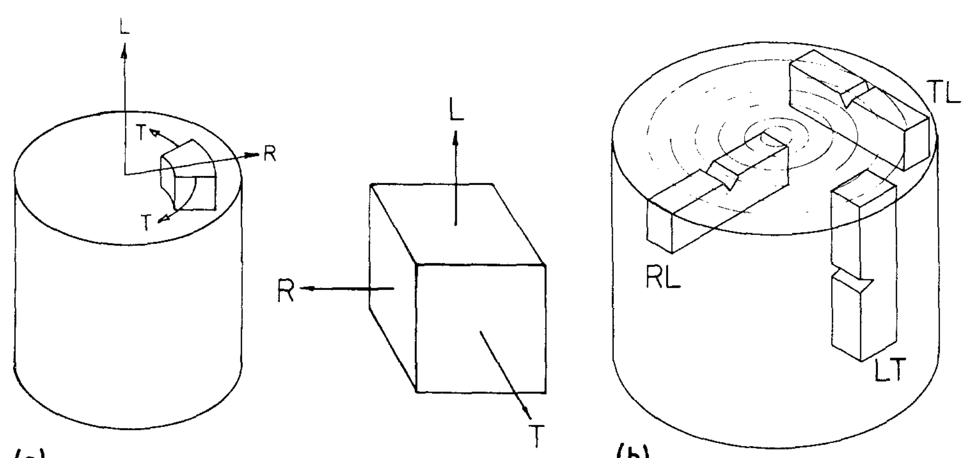


Figure 2 (a) The elastic planes of symmetry in wood. (b) The three principal crack propagation systems studied in this investigation.

Full experimental details are available elsewhere [8, 9]. The onset of crack extension was determined from the curve of load/crack opening displacement (measured using a double cantilever clip gauge mounted between aluminium knife edges across the crack). The "weak" TL and RL systems generally produced a distinct "pop-in" (sometimes coincident with an audible "click"), clearly indicating the point of crack initiation. With respect to the "tough" (LT) cracking mode, ultrasonic pulse velocity measurements [8] consistently showed a marked inflexion close to a 5% change in compliance (slope of the load–displacement curve). A "5% offset procedure", analogous to that described in the British Standards for Fracture Toughness Testing [10], was therefore used throughout for determining crack initiation in the LT system.

4. Microstructural influences on the fracture behaviour of wood

4.1. The effect of orientation

The nature of the orthotropicity of wood gives rise to weak planes parallel to the grain. Broadly speaking, the two so-called "tough" modes, LT and LR can have fracture toughness values up to an order of magnitude higher than the relatively weak, parallel-to-grain modes of crack propagation TL, RL, RT and TR [7, 11].

Fig. 3 shows the variation with grain angle of the fracture toughness, K_{Ic}. Grain angles of 0 and 90° represent the "weak" TL and "tough" LT systems, respectively. The curve shown is similar to the variation of the tensile strength with grain angle obtained by Baumann [12], and also to that obtained in studies on the tensile strength of unidirectional fibrous composites [13].

Tests also indicate that, regardless of the grain angle, macroscopic crack extension always occurs parallel to the grain. In the LT system, therefore, crack propagation takes place normal to the plane of the crack (Fig. 4) and is not immediately accompanied by complete failure of the specimen.

Although it is not appropriate here to comment in depth on the more general application of fracture mechanics to failure assessment in wood systems, it is of merit to comment briefly on the immediate implications of the "tough" mode of failure in wood.

Thus, the basic thermodynamic aspects of LEFM have been developed on the assumption that crack extension is self-similar, i.e. it occurs in the same plane as the crack, in a direction normal to the greatest tensile stress [14]. In particular, this is a necessary condition for the relationship to exist between the crack driving force and the stress intensity. Clearly, wood does not always satisfy this condition of crack extension, although a fractographic examination of LT specimens has shown that, under certain conditions, in-plane crack advance can occur from the initial crack tip for extensions of the order of the size of the cell, e.g. Fig. 5. However, this only satisfies the crack extension condition at a dimensional level at which wood is not a continuum, i.e. the linear elastic continuum to which the K_I values apply does not see the self-similar exten-

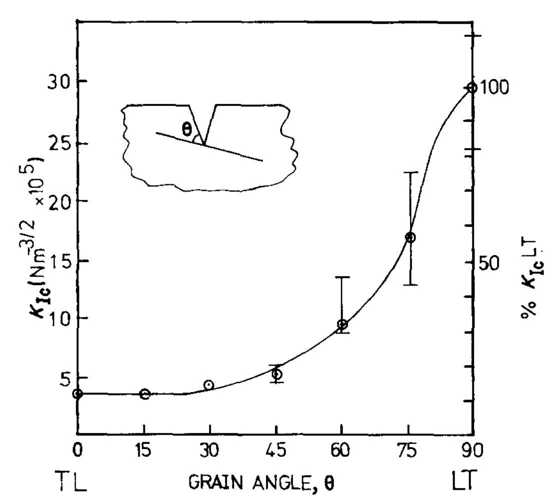


Figure 3 Experimental variation of fracture toughness, K_{Ic} with grain angle, θ.

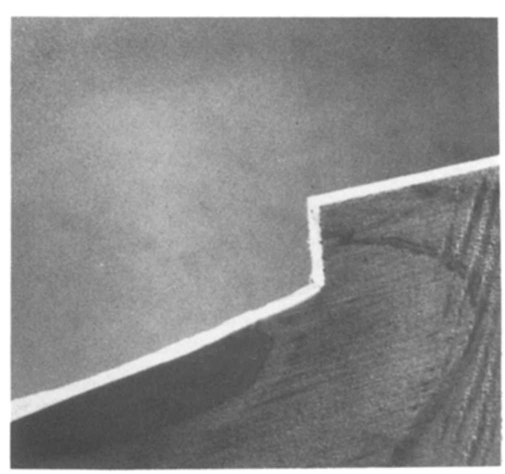


Figure 4 Crack propagation, normal to the plane of the pre-crack in the "tough" LT orientation, occurring parallel to the grain.

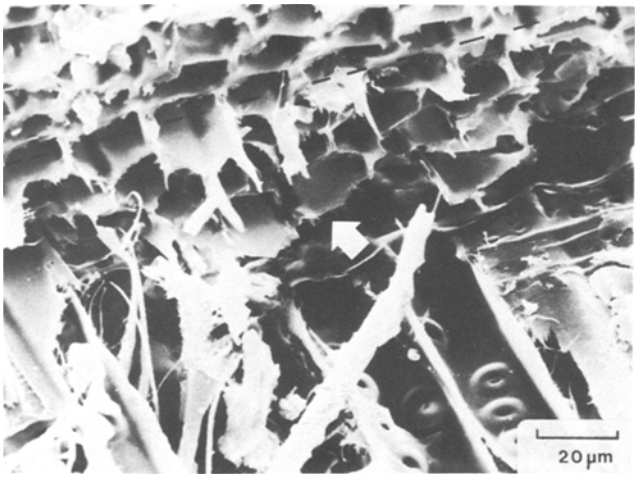


Figure 5 Local in-plane crack extension from a notch (dotted line) in the tough (LT) mode, involving the brittle, cross cell failure of large cells.

sion. Hence, the applicability of K_{Ic} to the tough LT mode remains in question and Schniewind's assumption [7, 15] that, in the tough systems, the crack is almost immediately diverted, does not entirely justify the analysis. (More recently, Barratt and Foschi [16] have, in fact, more realistically attempted to apply mode II stress intensity factors to the propagation of shear cracks in bending.)

Similar problems have been experienced in applying LEFM to artificial, unidirectional fibre composites, although Harrison [17] has shown through experiments on unidirectional glass reinforced plastic, notched perpendicular to the grain, that the effective value of K_{Ic} at the onset of splitting (i.e. shear crack propagation) was independent of the original crack length. This suggests, therefore, that although crack propagation is unlikely to be self-similar, fracture initiation may be consistent with a specific value of K_I. After crack propagation has occurred, however, K_I no longer applies to the new crack geometry; this, of course, immediately leads to the question that, if K_{Ic} does define the external load at which crack extension will occur in the LT system, whether or not this is really of any use in determining the stress level at which the specimen or structure as a whole fails. Indeed, whilst fracture mechanics parameters are undoubtedly useful for quantitatively categorizing material toughness (as used throughout this investigation), the more widespread application to realistic structural elements must be questioned [9].

Orientation effects are also in evidence under dynamic loading, or fatigue conditions. For example, the superior fatigue resistance of RL specimens, in comparison with TL specimens of the same thickness (Fig. 6), can be explained in terms of the effect of orientation on fracture mode. Thus, referring to the transverse section shown in Fig. 7, it can be seen that in the tangential direction the cells are slightly offset such that an RL system crack, in selecting the easiest fracture path, may cause both intercellular and intracellular failure, Fig. 8. The fracture resistance, under both static and fatigue conditions, will therefore be greater than for a corresponding TL system crack which generates relatively little new surface area in failing in a predominantly intercellular manner (depending on the cell size, see following). The inferior crack propagation resistance of the intercellular fracture path, for cells of the same size, is also clearly indicated by out-of-plane propagation observed in TL system specimens when the starter notch is not precisely parallel to the radial direction, schematically illustrated for clarity in Fig. 9.

Differences in crack propagation resistance between the RL and TL systems may also be explained in terms of the contribution of rays which, being perpendicular to the RL crack plane, are capable of fibre pull-out or buckling (Fig. 10a) involving much local damage and an increase in the work of fracture.

Ray cells also have an effect in the TL and LT systems by promoting cross-cell failure, under both static and fatigue conditions. Fig. 10b illustrates this common feature in TL fracture surfaces, and Fig. 10c shows the effect in the LT mode (also indicating signs of tension failure of the microfibrils, resulting in a stepped fracture, as arrowed). The cross-cell failure may arise due to

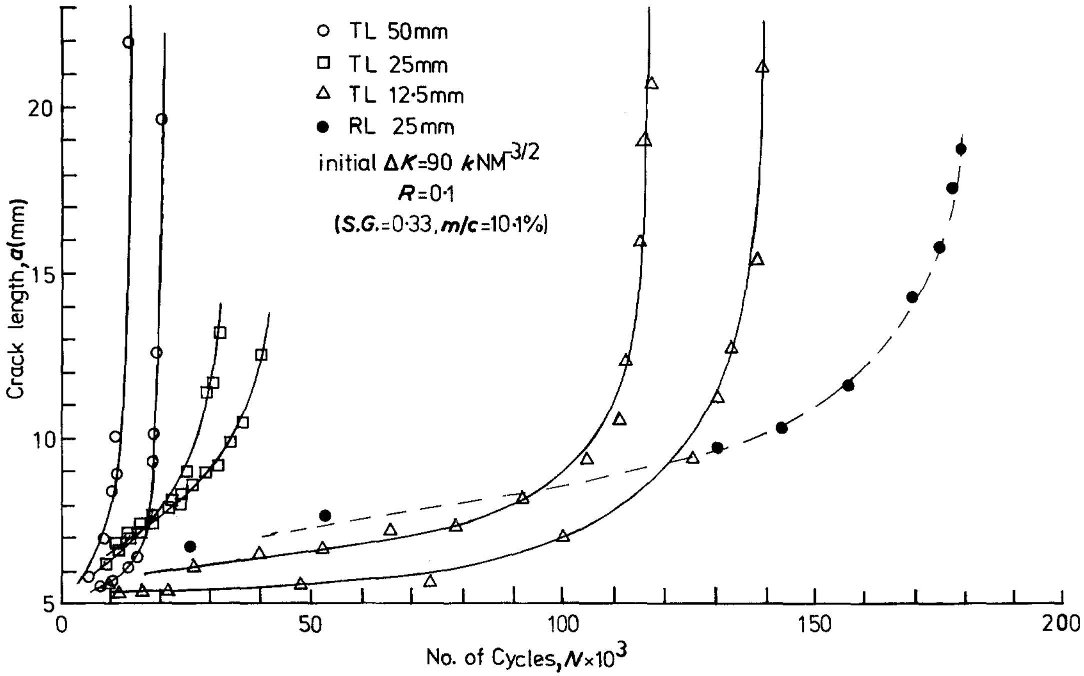


Figure 6 Fatigue crack growth curves illustrating the influence of specimen orientation on fatigue resistance, as well as the improvement observed with thinner specimens.

some structural weakening of the wall of the cell fibre as it traverses the ray.

Mechanical variability of individual cells may also be significant in affecting strength and toughness properties. For example, Furukawa and co-workers [18] have shown in tensile tests on individual cells that pits can initiate the tensile failure of the tracheid. Fig. 11 shows an area of fracture around a pit in the tough LT fracture system, from which it it evident that the junction of the pit with the cell wall does seem to constitute a tensile weakness in the cellular structure. Densely pitted areas might therefore be expected to have lower local strength than generally expected from the tracheid.

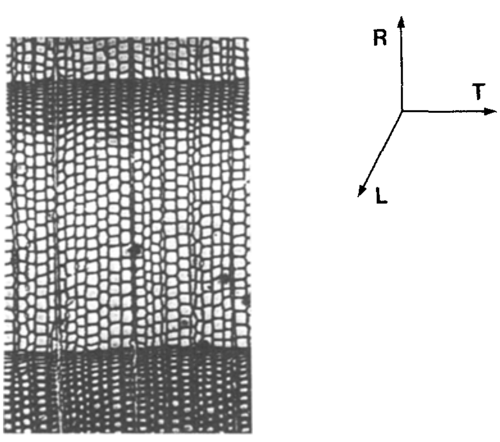


Figure 7 Transverse section of softwood in relation to radial, longitudinal and tangential directions (from Kollman and Coté, [12]).

4.2. The effect of cell size

Previous workers [19–21] have found that weak-mode fracture surfaces can be conveniently characterized with respect to the failure mode, i.e. intercellular or intracellular, Fig. 8a.

Debaise and co-workers [19, 21] have maintained that, for TL specimens in the forward shear and simple opening modes, slow crack growth is largely intercellular, while rapid failure occurs in an intracellular manner. The present study, however, has indicated, in accord with the review by Dinwoodie [1], that microstructural variations in cell size play a dominant role in determining the fracture mode. Thus, Fig. 12a shows the boundary between small and large cell failure of TL specimens during a static test (fatigue fractures producing identical effects). It is immediately evident that the small cells on the left have failed in an entirely intercellular manner, compared with the larger cells which have fractured in an intra- (or trans-) cellular mode, exposing the cell lumens in which pits can clearly be seen.

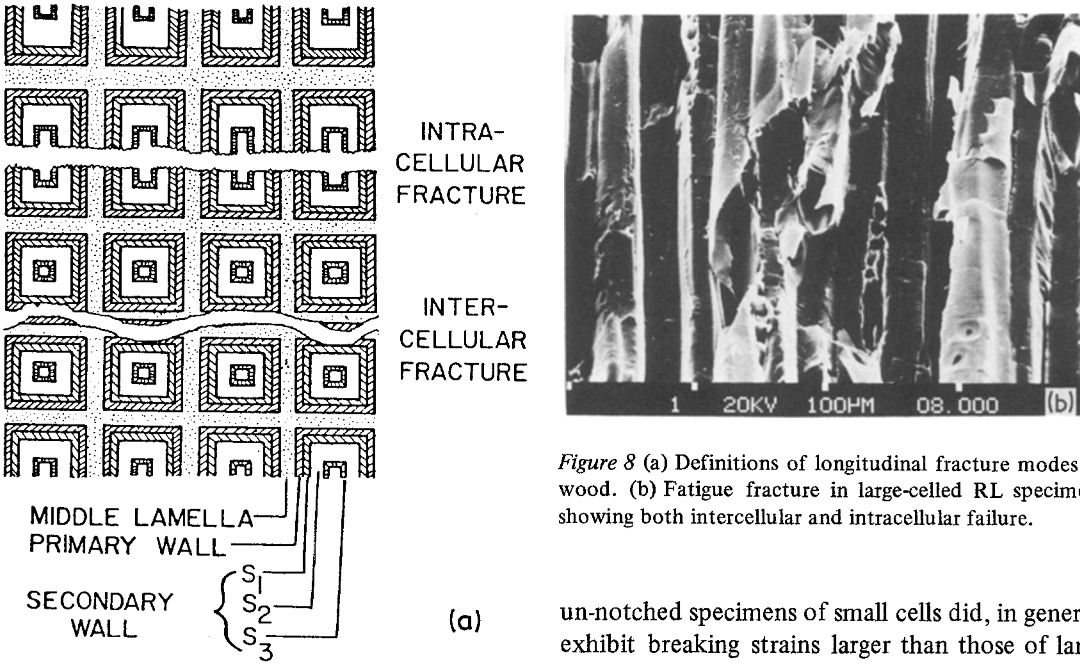


Figure 8 (a) Definitions of longitudinal fracture modes in wood. (b) Fatigue fracture in large-celled RL specimen, showing both intercellular and intracellular failure.

Fig. 12b shows at higher magnification an area of intercellular failure for a small-celled region of a TL specimen. In the area of separation between the tracheid cells, fracture has occurred in the region of the middle lamella and the primary wall, in general agreement with the observations of Koran [20] in un-notched TL specimens. At the bulge of the cell, failure appears to have extended into the cell wall to reveal the shallow helical windings of the S_1 layer. In other areas of intercellular failure, particularly close to the small cell–large cell boundary, a transition zone may be observed where fracture extends well into the middle cell wall (or S_2) layer, Fig. 12c.

In the large cells, having thinner walls, fracture across the cell walls presumably presents the easiest fracture path, although significant deformation typically accompanies such intracellular failure (Fig. 10b). Thus, thin specimens of the TL system containing only large cells (and showing mainly intracellular failure) have lower K_{Ic} values than those containing small cells only (see Fig. 13a), where fracture was predominantly intercellular. Corresponding differences in fracture mode resulted from thin specimen tests on small and large celled samples of the RL system, but with a surprisingly small difference in toughness, Fig. 13c.

The effect of cell size on toughness is consistent with experiments [22, 23] which have shown that un-notched specimens of small cells did, in general, exhibit breaking strains larger than those of large cells, in contrast to individual fibre load curves [24] indicating that large cells are capable of much greater deformations prior to fracture. Ifju [23] has interpreted this paradox with reference to the difference in fracture mode, by suggesting that intercellular failure of the small-celled specimens is equivalent to fibre-pullout, thus enhancing composite ductility as well as work of fracture.

In concluding this particular discussion, it is of interest to point out that increasing proportions of the "tougher" small cells in a test piece cross-section could be construed as an explanation for increasing fracture toughness with diminishing specimen thickness, particularly in the TL system

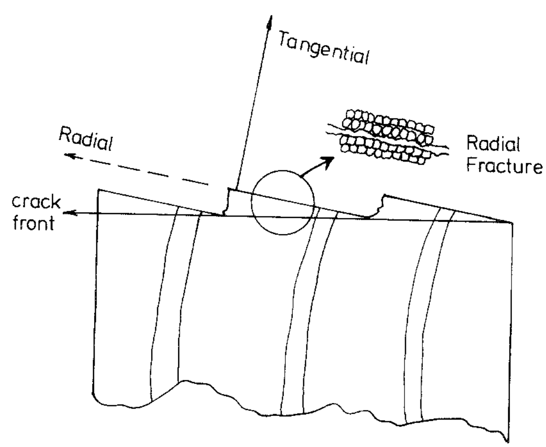
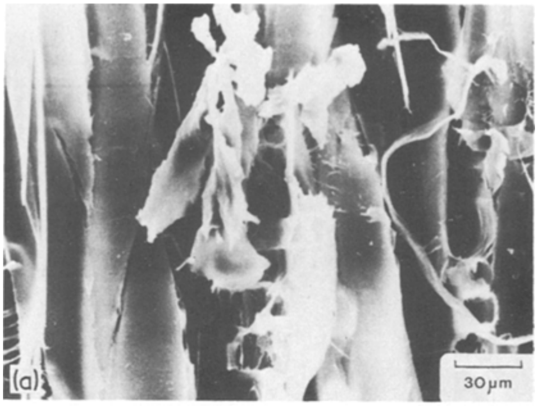
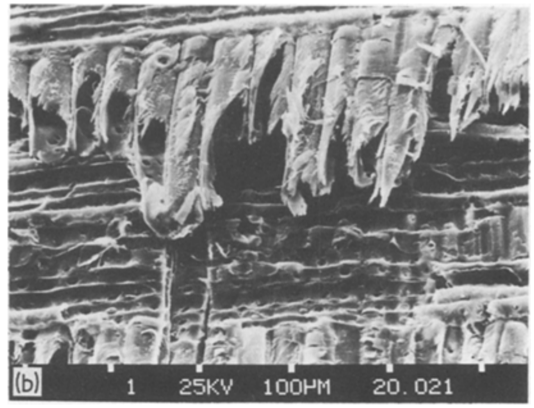


Figure 9 Fracture path observed in the specimens with 15° grain angle, in which crack propagation occurs out of the TL notch plane, but along a true radial direction.





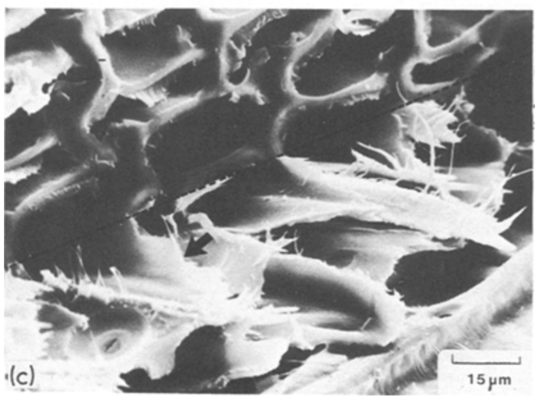


Figure 10 (a) RL fracture surface of large cells showing extensive local damage associated with the failure of ray cells. (b) Cross cell failure exhibited by large cells near rays in the TL system (fatigue failure in this case). (c) Cross cell fracture from a notch (as marked) in the LT mode, with stepped cell wall fracture arrowed.

(Fig. 13a). However, with regard to the 3 mm specimens in the LT system, these samples consisted entirely of large cells and had fracture toughness values of about twice that of the thicker specimens containing some small cells as well as both the 1 mm specimens (large and small cells only). Coupled with the limited effect of cell size on specimen toughness in the RL systems mentioned earlier, these observations indicate that the higher toughness values at reduced thickness are not due to microstructural factors. The nature of this effect will be discussed later in the paper.

5. The influence of stress state on fracture in wood

5.1. The evidence for crack front bowing

At an early stage in the experimental investigations it was evident, both from static and fatigue tests on pre-cracked samples, that fracture initiation occurred first at the centre of test specimens. The preliminary indication of this was when the onset of fracture was distinctly audible, but no cracking was visually apparent on either side of a test piece.

As a follow up to this qualitative observation, a series of fracture toughness tests were stopped immediately after crack initiation had occurred, as indicated by a characteristic load curve "pop-in" and confirmed by ultrasonic pulse transit time measurements [8]. The specimens were removed from the test rig and ink was squirted into the crack with a syringe, taking great care to ensure a uniform distribution of ink across the crack front. Each specimen was lightly stressed manually to allow the ink to seep into the crack. (Control tests on specimens "cracked" with a razor blade cut confirmed that seepage past the crack front occurred uniformly and to a depth in the region of 1 to 2 mm, depending on density

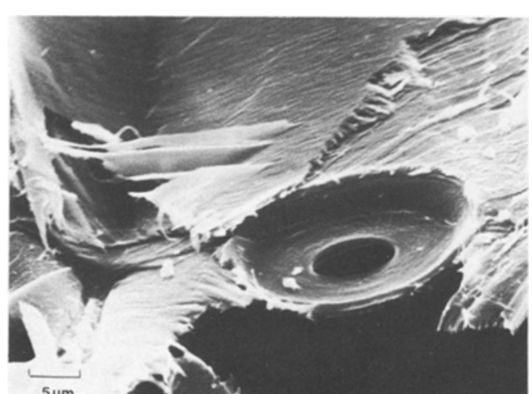
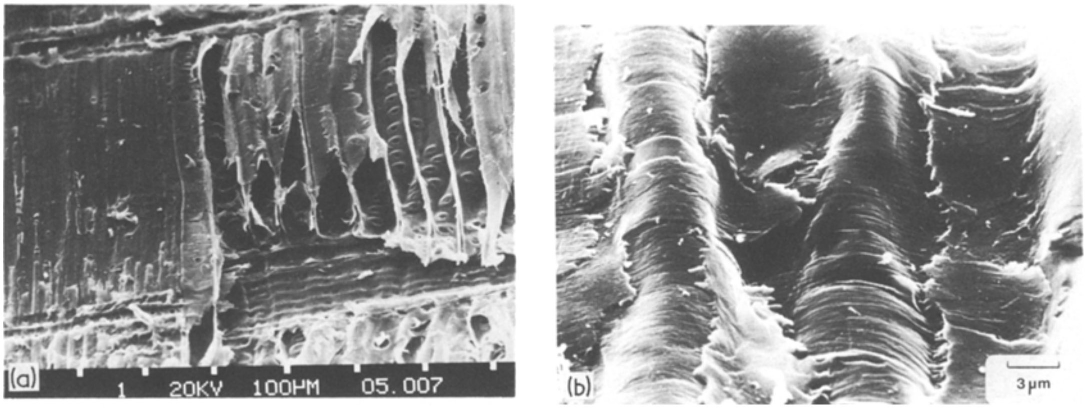


Figure 11 Fracture of a pit in the large cells of an LT specimen; the pit itself is undamaged, and failure has occurred at the pit-cell wall junction.



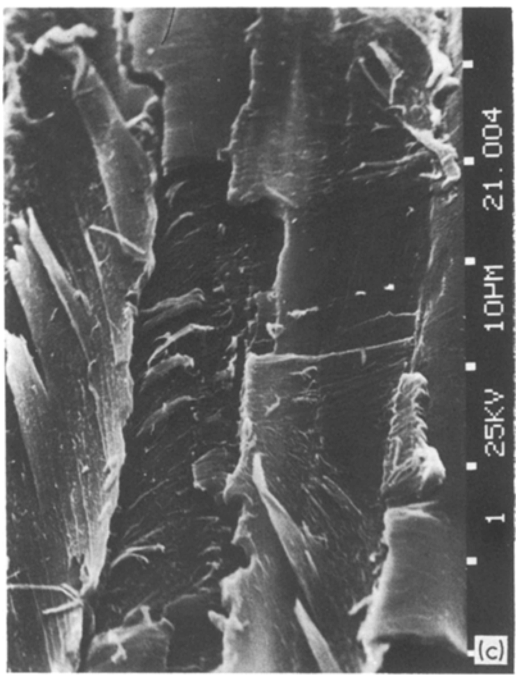


Figure 12 (a) Intercellular failure of small cells on the left, bounding on to a region of large cells which have failed in a largely intracellular mode. (b) Higher magnification fractograph of intercellular fracture in small cells, TL system. (c) Exposed middle cell wall in small cell fracture close to a small cell-large cell boundary.

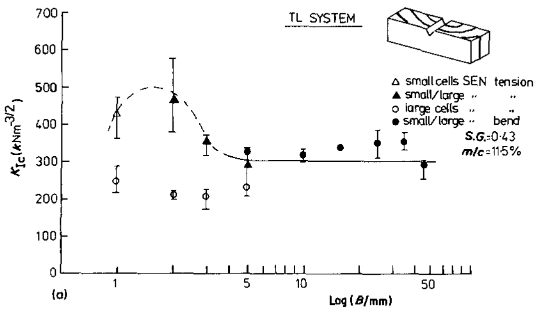
variations with wood growth patterns.) On breaking open the specimen, the position of the crack front was clearly delineated. Preferential crack bowing was conclusively demonstrated for both the TL and LT orientations, Fig. 14, where the initiation of cracking was stable under stroke-controlled conditions, and at five grain angles between the two systems. In the RL system, crack initiation leads almost immediately to general failure so this ink stain method was not suitable.

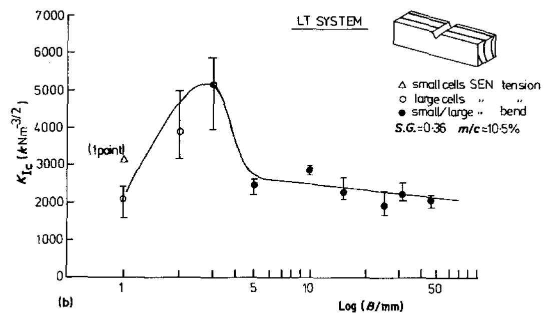
A repeated observation was that crack tunnelling in both the TL and LT systems occurred not in the precise centre of the specimen, but slightly nearer the edge furthest from the pith (Figs. 14b and c), i.e. closer to the bark side. This effect appears to be due to the bands of small-celled wood which tend to be closer together (and therefore cause a local increase in density) the further they are away from the pith. This conclusion draws support from the observation that in denser wood, containing more evenly spaced growth rings, cracking initiated at the centre rather than towards one edge.

Certainly these observations, which indicate fracture to initiate preferentially (and presumably, therefore, more readily) from the centre of the crack front, suggest that fracture in wood is dependent on the state of stress ahead of the crack. It is of merit, therefore, to examine in more detail both the distribution of stresses ahead of a crack under load in wood as well as other experimental implications of this conclusion.

5.2. The crack tip stress/strain field in wood

With appropriate adjustments for orthotropicity (specifically the effect of elastic anisotropy on Poisson's ratio), the major features of the elastic isotropic stress field distribution ahead of a crack can reasonably be applied to wood. For example, under plane strain conditions transverse stresses ahead of the crack will be higher for the "tough" LT condition than for the corresponding isotropic case, whereas in the RL and TL systems (having wood cells parallel to the direction of crack-propagation) transverse stresses are significantly reduced [8].





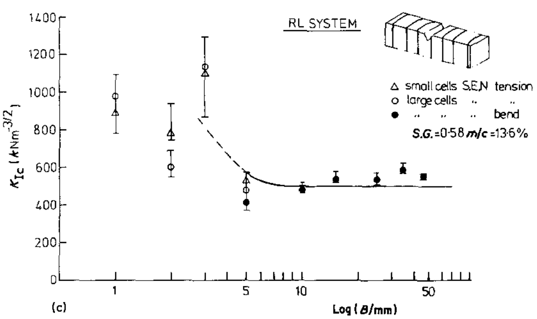


Figure 13 Variation of fracture toughness, K_{Ic} with specimen thickness, B for (a) TL, (b) LT and (c) RL crack propagation systems.

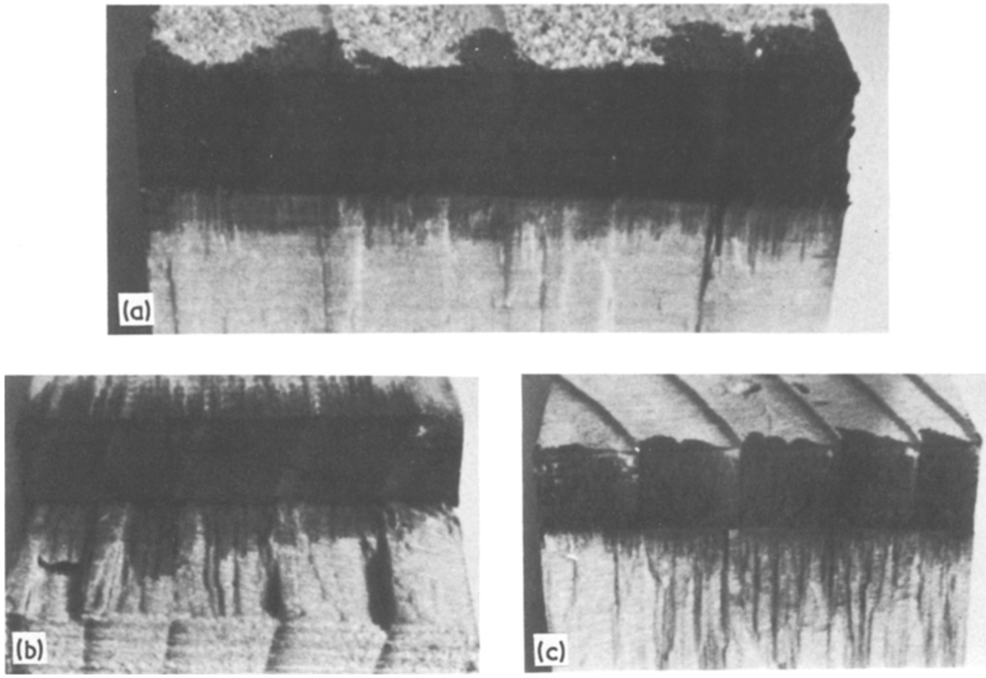
Where a material can exhibit permanent deformation (as is the case with wood), the elastic stress distribution at the crack tip will be altered, depending on the prevailing stress state. Thus, under conditions of crack tip triaxial stress (plane strain), the material in the crack tip "process zone" will be at a higher stress than in the equivalent plane stress condition (where there is no through-thickness stress and therefore no stress triaxiality ahead of the crack tip) where yield will occur at the normal uniaxial yield stress. In addition, under plane strain conditions the region where the local stress distribution exceeds the local yield stress is confined to a smaller distance from the crack tip than in the plane stress state.

Although in the first instance it is somewhat difficult to imagine any precisely analogous process to the plastic flow of metals occurring in wood, it seems plausible that some deformation zone can exist at the tip of any fibrous composite, including wood. Indeed, Kanninen and co-workers [25] have approached the problem of failure in composites by considering the material to consist of two regions. The region outside the immediate vicinity of the crack tip is still regarded as an orthotropic continuum, while the material near the crack tip is considered as being "real". Fracture processes in this local, heterogeneous region are then considered in detail and a failure criterion is developed by relating the energy available in the stressed continuum to the energy needed for rupture in the local crack tip region.

The deformation zone would be subject to similar variations in stresses and strains with the thickness of the specimen, as is the so-called plastic zone in metals, and be dependent as such regions are on the particular states of stress at the crack tip. It is useful, therefore, to employ the convention normally applied to metals in describing the region at a crack tip where the material has undergone permanent deformation as a "plastic zone". Thus, to reiterate, a plane strain crack tip region will be subjected to relatively higher stresses, both within the plastic zone itself and just ahead of it. Furthermore, the plane strain plastic zone will give rise to a greater strain gradient ahead of a crack subjected to load, since the strain concentration in the crack tip zone is greater than for the plane stress condition.

Therefore, whether fracture is promoted by the higher tensile elastic stresses ahead of the plastic zone, or by the larger strains in the plastic zone itself, preferential fracture will occur in the plane strain region; crack extension, therefore, will be expected to take place first in the centre of thick specimens, as observed.

In order to examine experimentally this conclusion, an attempt was made to measure the strains ahead of a crack tip. Two four-ply, notched glued laminated specimens were prepared, in the LT and TL systems, with strain gauges positioned across the crack front, ahead of the crack tip. For the TL specimen the strain gauges were placed to within 1 mm of a razor cut notch, while the LT notch was left rounded and the gauge grid was placed approximately 2 mm from the notch tip.



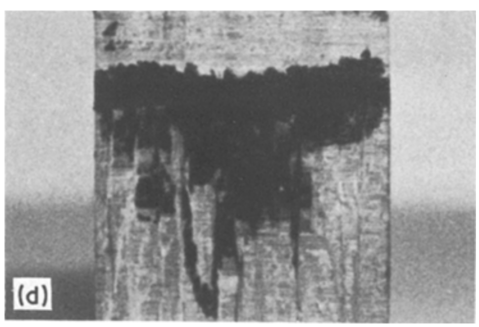


Figure 14 Inked crack fronts and the evidence for crack front bowing: (a) TL control specimen (uncracked) showing seepage past the razor cut pre-notch; (b) LT and (c) TL specimens, static loading and (d) TL, fatigue loaded.

The specimens were loaded in four-point bending and the corresponding strains are shown in Fig. 15. Clearly, the strains in the centre of the testpiece are consistently higher than those at the edge, indicating that the longitudinal elastic tensile strain field ahead of the crack varies along the crack front, in a manner consistent with the previous discussion.

An interesting result, evident in Fig. 15, is that the strains at the actual centre of the specimens are uniformly lower than those somewhat closer to the edges of the specimen. This is rather unexpected and cannot be adequately interpreted at this point in time. Certainly fractographic examination (described in some detail as follows) has not revealed any mechanistic basis for this effect, experimentally consistent though it is.

5.3. Thickness effects on toughness and fatigue crack propagation

A further manifestation of stress state variations on fracture characteristics, where such variations can be shown to influence material deformation behaviour, is the effect of specimen thickness.

A number of studies, recently reviewed [26], have indicated that thicker specimens have inferior fracture properties compared to thin ones. Often this may be associated with the fact that large members are not normally defect free, but similar trends have been reported when comparing clear specimens of varying thickness [26].

In the context of the current investigation, the effect of thickness on both "static" fracture toughness and fatigue crack propagation rate was examined. In the first instance, pre-cracked TL, RL and LT specimens of thickness varying between 1 and 45 mm were tested to determine K_{Ic} values according to the methods described in Section 3. The results (shown in Fig. 13) firstly confirm that thinner specimens have a higher toughness, and secondly, indicate that the toughness levels in thicker specimens achieve a lower limiting value

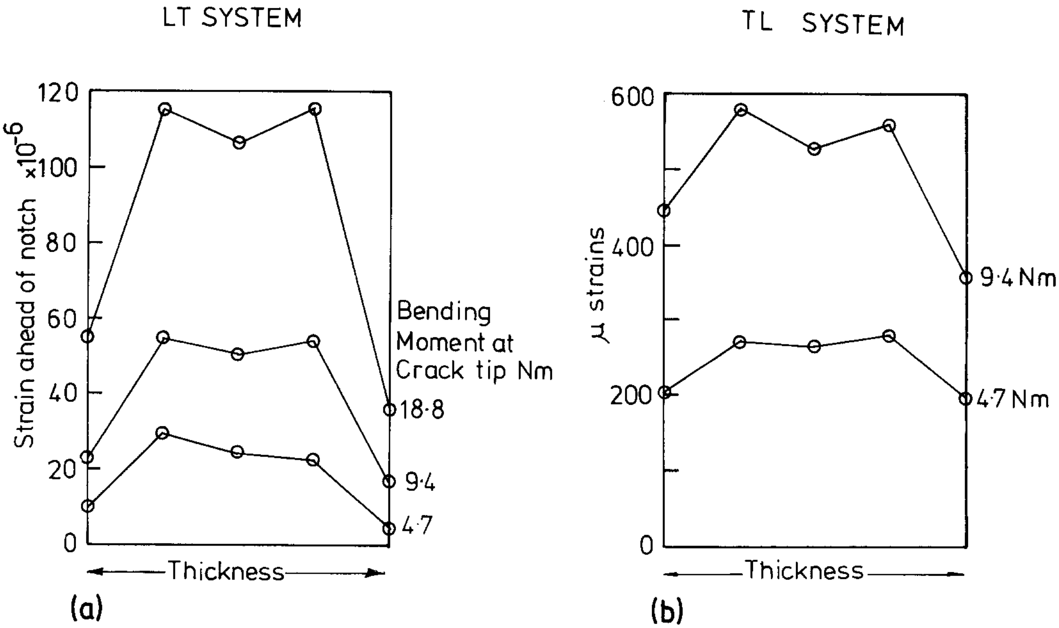


Figure 15 Variation with specimen thickness of longitudinal strain ahead of a notch, as measured experimentally in a fabricated laminate, for various bending moments at the notch plane.

with increasing thickness. Further, there is some evidence to suggest that in very thin specimens toughness falls off with decreasing thickness, as occurs in metals where a stress state dependent deformation mode occurs.

Under fatigue conditions, a similar influence of specimen thickness was observed, Fig. 6. Here, prenotched specimens were tested in bending at 20 Hz, with an initial ΔK value of 90 $kNm^{-3/2}$ and an R value ($= K_{min}/K_{max}$) of 0.1. Crack monitoring was carried out by surface measurements only, although ink staining studies (described earlier) showed the crack front to be irregular but typically bowed at the centre (Fig. 14d). Again, these results show that thinner specimens have greater resistance to cracking. Also, under fatigue conditions the results indicate that the effect of thickness persists to higher values than for the corresponding static toughness tests.

While there have been some efforts in the literature to interpret the thickness effect on fracture properties in wood on a statistical basis developed from a Weibull-type analysis [27], Boatright and Garrett [28] have recently shown that this approach cannot be justified from a number of points of view. A major consideration concerns effects described in this paper, where cracks have been observed to tunnel into the specimens, under both static and dynamic loading, leaving uncracked ligaments at the edges which in due course are broken as the crack propagates further. Fracture in this case cannot be solely dependent on local strength inequalities of the volume elements ahead of the crack tip (a fundamental precept of any statistical analysis), simply because there is no reason whatsoever for the weakest elements to occur always at the specimen centre.

In summary, therefore, crack bowing observations, coupled with direct crack tip strain measurements and a verification of the influence of specimen thickness on fracture properties, would appear to confirm the existence of stress state variations across a crack front in wood and a corresponding influence on fracture behaviour. There remains, however, the need to clarify the nature of the permanent deformation which must be occurring in the high stress regions of a crack tip, and the manner in which this deformation is affected by stress triaxiality.

5.4. Stress state and deformation mode

Scanning electron fractography was carried out on both LT and RL specimens in order to determine whether or not the observed toughness variations with specimen thickness were accompanied by any change in fracture mode.

In thin specimens, and close to the edges of thicker specimens, deformation modes involving cell debonding, axial twisting and buckling were

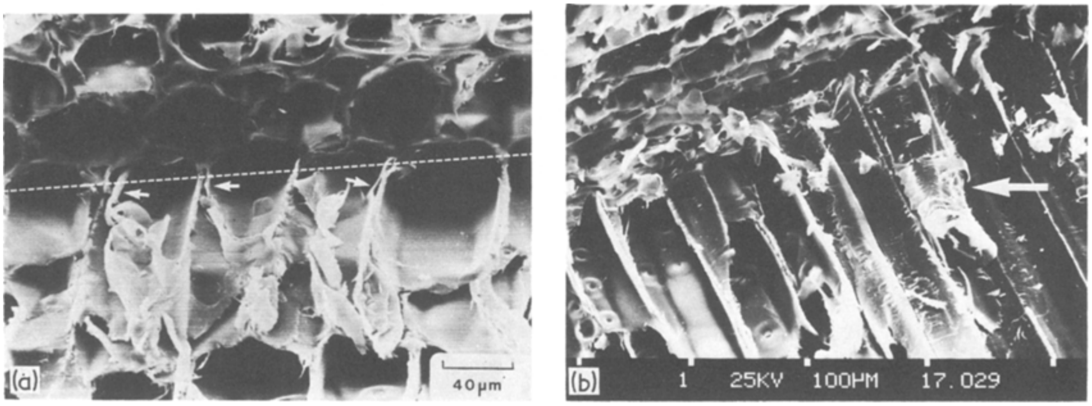
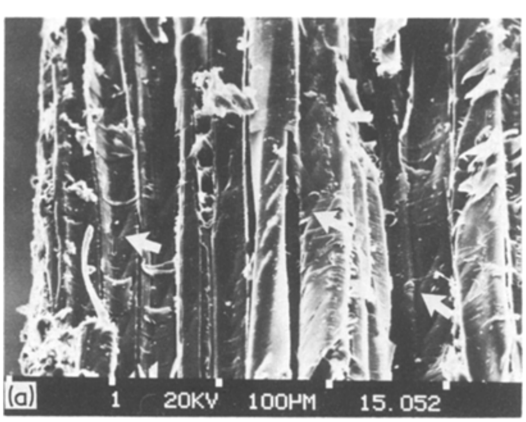


Figure 16 Large-celled, LT orientation specimens showing (a) cell debonding (arrowed) and (b) buckling deformation (arrowed), at the notch tip in, respectively, 1 and 3 mm thick sections.

observed that did not seem to be present in the central regions of the thicker specimens. Thus, Figs. 16a and b show evidence for cell debonding and deformation in, respectively, 1 mm and 3 mm thick LT ("tough" mode) specimens. In Fig. 16a the cell walls have evidently failed either at the interface between the middle lamella and the primary wall, or at the S_2/S_1 boundary. This debonding has been accompanied by extended deformation prior to failure, as indicated by the distended and fibrous nature of the remaining portion of the cell wall.

A similar deformation mode was observed in thin RL specimens. Thus Fig. 17 shows a typical fracture surface consisting of small cells in a 1 mm thick test piece, from which it is immediately apparent, firstly, that failure is almost entirely intercellular (evidently in and around the region of the middle lamella) and secondly, that extensive shear deformation has taken place, probably in the secondary layer of the cell wall. One of the cells shown in the centre section of Fig. 17a has clearly become separated and around this cell extensive debonding has occurred at the primary/ S_1 interface, indicated by the shallow angle of the microfibrils visible at the outer layer of the cell shown in Fig. 17b. In contrast, the lines of shear are much steeper and indicate failure in the S_2 layer, which has also caused shear failure in the



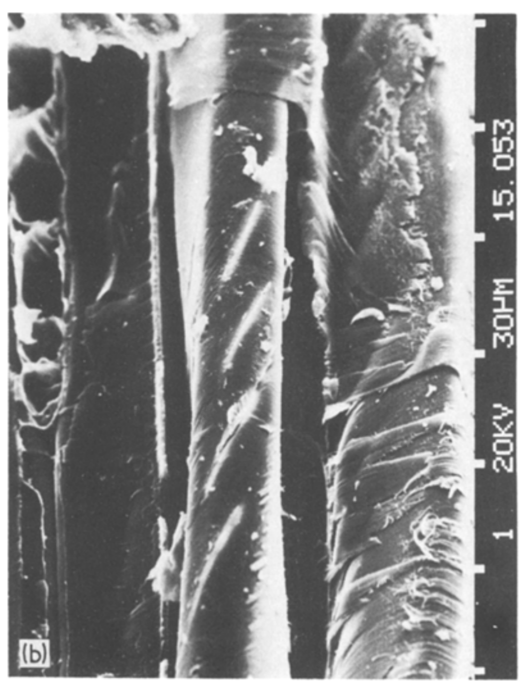


Figure 17 Fracture of RL specimen, small cells, 1 mm thick testpiece, showing (a) extensive call wall shear, as arrowed, and (b) at higher magnification a debonded cell showing microfibril angle in S_1 layer and angle of apparent shear in the S_2 layer.

primary layer, not aligned with the microfibril direction in this layer.

Although the possibility exists that the lines of shear are broadly aligned with the S_2 microfibrils, detailed consideration *in situ* in the scanning electron microscope (SEM) have tended to indicate that they cannot readily be attributed to any anatomical feature of the wood under test. In any event, the scale of the deformation under consideration appears to derive from through-thickness relaxation. Thus, under plane stress conditions, planes of maximum shear in the RL and TL systems will be parallel to the cell axis; there will, therefore, be an increase in the shear stress acting intercellularly in the RL system in this stress state. This may induce intercellular failure which in turn may contribute to the through-thickness strain. Lateral contraction of the cells under such conditions could reasonably be expected to promote the type of shear failure observed.

Shear markings could not be found on the fracture surface of thicker specimens, except near the specimen edge. Fig. 18a shows shear lines close to the edge of a 5 mm RL specimen (having a nominal fracture toughness equivalent to that found at all greater thicknesses, Fig. 13c). A higher magnification fractograph, Fig. 18b, illustrates that the shear fold is clearly not in the same direction as the microfibrils in what is apparently the primary wall, although it is difficult to eliminate unequivocally the possibility that the deformation is aligned with the microfibrils in the S_2 layer. Similar cell deformation was also observed under fatigue conditions (and indeed fracture morphology under both static and fatigue conditions was remarkably similar [8]). Thus, Fig. 19a shows the buckling deformation of large cells in the RL system while Fig. 19b gives a close-up of shear failure of the S_2 layer, behind the visible S_1 layer.

The cell buckling reported above has a clear similarity to the type of axial buckling producing a longitudinal folding of the cell wall first reported by Page and co-workers [29] from tension tests on single wood fibres. However, the cell deformation observed in this current investigation and elsewhere [30] is unlikely to have been caused by axial tensile stresses. As mentioned earlier, due to the low Poisson's ratio for contraction in the longitudinal (axial) direction which is induced by lateral strains, the transverse stresses ahead of the crack can be expected to be relatively small. The twisting of the cell is likely to have been caused by shear stresses induced during fracture, as already noted. The essential observation, however, is that the kind of irreversible deformation observed must be contributing to the recorded higher toughness of the thin specimens, as well as to the retardation of crack advance at the edges of thicker specimens. Indeed, on this basis it is of interest to note that Gordon and Jeronomidis [31] have already recognised cell buckling as a mechanism of absorbing strain energy, using it to calculate a work of fracture for wood in reasonable agreement with experimental values. Furthermore, the mechanism forms the basis of the design of a new artificial composite with an exceptionally high work of fracture [32].

In concluding this particular discussion it must be emphasized that arguments relating to the local deformation behaviour of bulk wood samples are still incomplete since undoubtedly, as a whole, the process is extremely complicated. Permanent

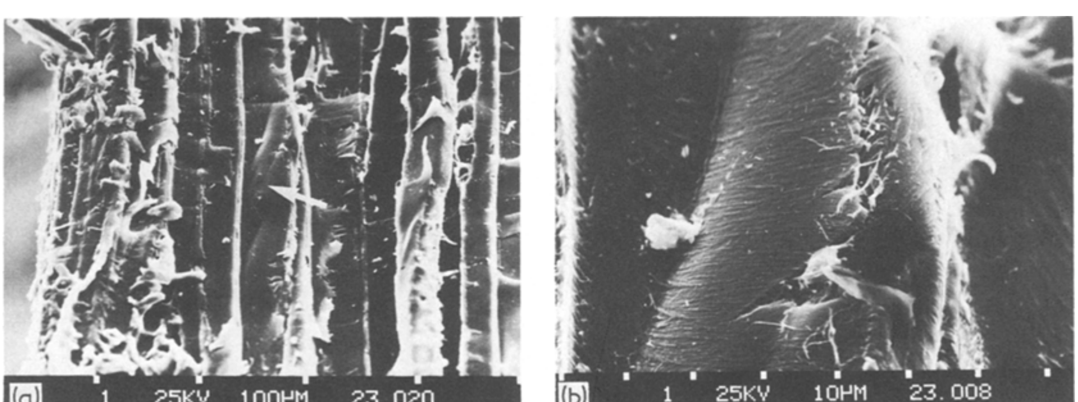


Figure 18 (a) Shear lines, as arrowed, near the edge of a 5 mm thick RL specimen. (b) Higher magnification fractograph of central region of (a), illustrating clearly that the shear fold is not aligned with the horizontally aligned microfibrils.

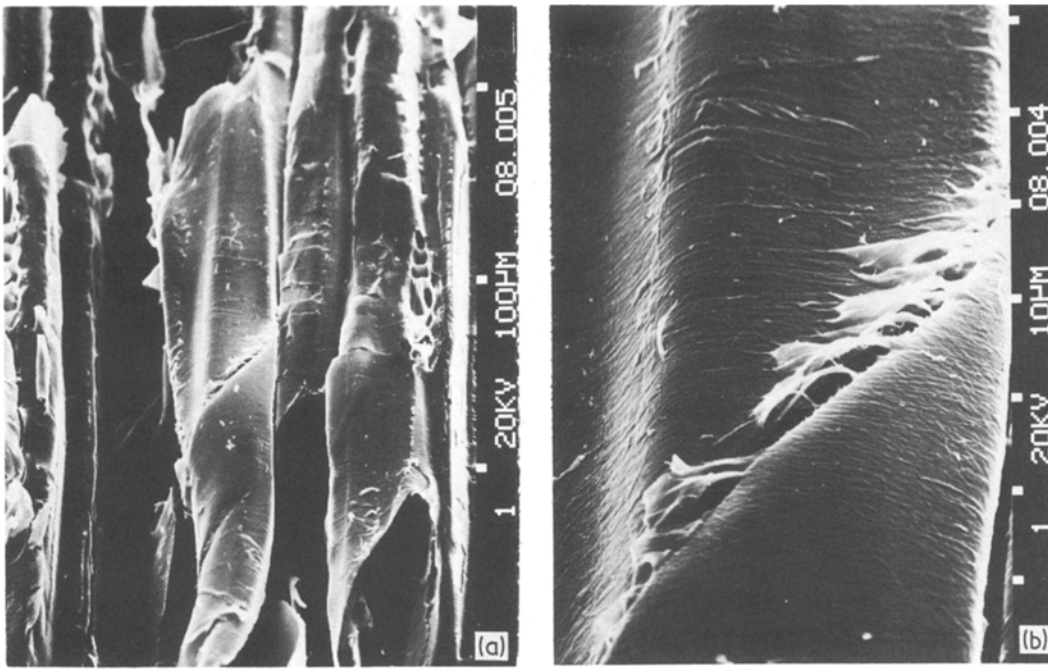


Figure 19 (a) Buckling deformation during fatigue fracture of large cells in the RL system. (b) Higher magnification fractograph of the shear failure of the S_2 layer, behind the visible S_1 layer.

distortion of the cellular structure may involve fracture (cell debonding, and shear within the cell wall), micro-plasticity of the cell components as well as deformation of the cell unit itself (buckling). Certainly, various previously reported elastic models of the cell and cell wall (e.g. Mark in [33], and [34–36]) have not, as yet, been successfully extended to analyse post-elastic behaviour.

5.5. The effect of notches on fracture mode

Tests have indicated that notches substantially influence the mode of failure of "tough" LT wood specimens. Fatigue experiments on pre-notched test pieces, for example, showed crack propagation to take place from the crack tip parallel to the grain. Other fatigue results [37, 38] in the presence of artificial or natural checks also report parallel-to-grain failure involving the propagation of a major crack and fracture of the specimen by shear. In contrast, un-notched bend fatigue failures (e.g. [39]) describe a failure mode more closely akin to normal, "splintery" fracture, i.e. where failure has occurred at a number of independent interfaces within the bulk of the material. This difference in behaviour between notched and plain specimens, tested under slow (static) bending, is clearly illustrated by comparing Fig. 4 with Fig. 20. In the un-notched specimen, Fig. 20, progressive failure occurs over a diverse area of the stressed section, and the points of local damage finally coalesce to cause gross "cumulative" failure. A corresponding notched specimen, however, fails ostensibly in a non-cumulative manner (Fig. 4), caused by the propagation of a single, major crack.

The observed difference between notched and un-notched specimens can be attributed to the

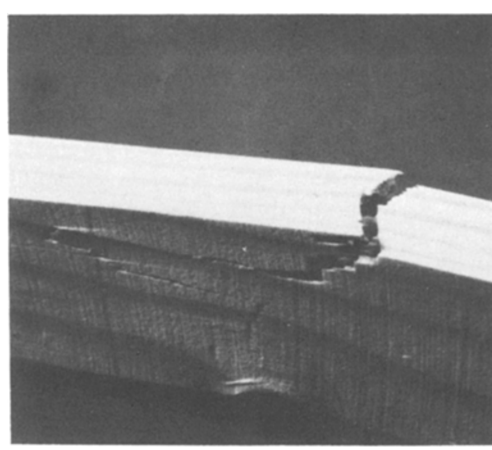


Figure 20 Cumulative failure of an un-notched "tough mode" (LT) sample tested in bending.

respective distribution of shear stresses (required for delamination) in the two cases. Thus, in a notched sample the maximum shear stress (and tensile stress) occurs in the crack tip region. In an un-notched beam, however, with the maximum shear stress at the centre of the beam, conditions are favourable for crack initiation but, because of the zero tensile stresses along the central neutral axis, correspondingly unfavourable for subsequent opening and propagation of an interfacial crack. Progressive, cumulative failure, therefore, results from a compromise between local shear and tensile stress distributions and the manner in which these are affected by cracking.

Failure in tough mode (LT) samples containing knots has also been observed to be cumulative, and this difference in failure mode between notched and plain specimens, in bending at least, is considered to be a major reason why a fracture mechanics treatment cannot be realistically used for the strength assessment of knotted timber [9].

6. A strain-controlled criterion for fracture in wood

A macroscopic strain criterion for fracture may be applicable when some degree of ductility is exhibited in the conventional stress–strain curve; fracture is then no longer uniquely defined by a specific stress, but rather by the attainment of a sufficient strain. Evidence leading to a possible strain-controlled fracture criterion in wood is, therefore, rather unexpected, since the gross tensile behaviour of wood shows little or no ductility, and is usually regarded, at least in the engineering sense, as being brittle. However, as pointed out by Gordon and Jeronomidis [31], the high work of fracture of wood perpendicular to the grain is hardly compatible with the observed brittle gross tensile behaviour. Certainly energy absorbing fracture mechanisms such as the cell buckling mode described in a previous section can offer a rational interpretation of any strain-controlled fracture mechanism, so the concept is worth pursuing.

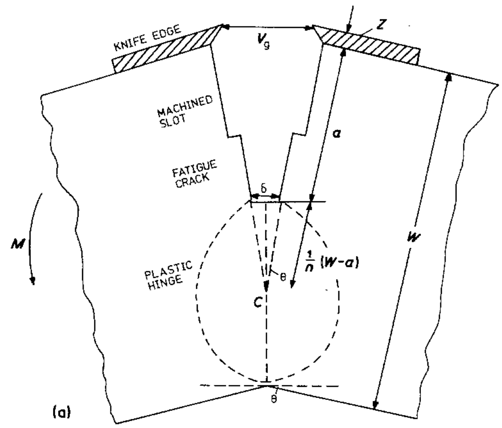
Nearly twenty years ago, research [22, 23] had shown that the ultimate tensile breaking strain of small, thin specimens, somewhat larger than individual fibres, is independent of the variables which usually influence the mechanical properties of wood (such as microfibril angle and moisture content). A further interesting result from the early literature [40] indicates that, although an increase in the rate of loading in bending produces a corresponding increase in both the limit of proportionality and the ultimate bending strength, the central deflection (i.e. the strain) at failure was essentially constant, regardless of the loading rate. This is significant in so far as it suggests that a strain-controlled fracture criterion could be applicable, irrespective of visco-elastic deformation occurring within wood.

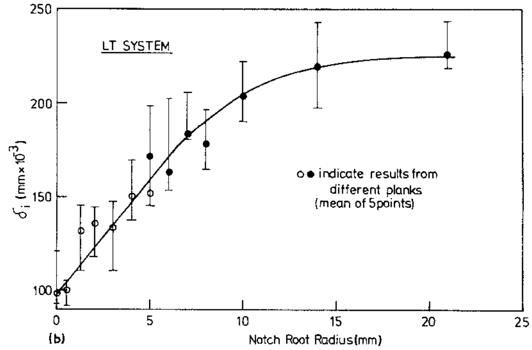
Based on Jayne's results [24], which showed that individual small cells and large cells have similar breaking strains when pulled axially, Perkins [41] has previously postulated a strain criterion, suggesting that the fracture process may initiate when the strain in some region of the cell wall reaches a critical value. (It should be noted, however, that Jayne himself observed that there was a trend for higher breaking strains in the large-celled wood [24].)

In order to examine the validity of a strain-controlled criterion for the initiation of fracture in pre-notched testpieces, crack opening displacement (COD) measurements were made at the onset of cracking in bend specimens. Crack tip strain can be related to the COD, and the COD has been used with success as a fracture toughness parameter in metals and alloys [42], specifically where high toughness materials are concerned when linear elastic fracture mechanics is inapplicable. Essentially, crack extension can take place when the material at a crack tip has reached a maximum permissible strain, corresponding to a critical value of the COD.

For two crack plane orientations (TL and LT) COD tests were carried out on specimens having identical crack lengths, but with varying notch root radii. (The notches were inserted by first drilling a hole at the required crack length, and then cutting back to the hole with a hacksaw.) Values of the crack tip opening displacement at fracture initiation, δ_i, were calculated conventionally (e.g. [42]) and the results are shown in Fig. 21. In the TL system, δ_i increases linearly with notch root radius, r; in the LT system, specimens of larger hole sizes were also tested and the initial linear region shown levels off to a constant δ_i for a notch radius in excess of about 15 mm.

The achievement of a limiting constant δ_i value at higher notch root radii was accompanied by failure tending to initiate away from the apex of the hole. This trend can be explained in terms of the results of Green and Taylor [43], who showed





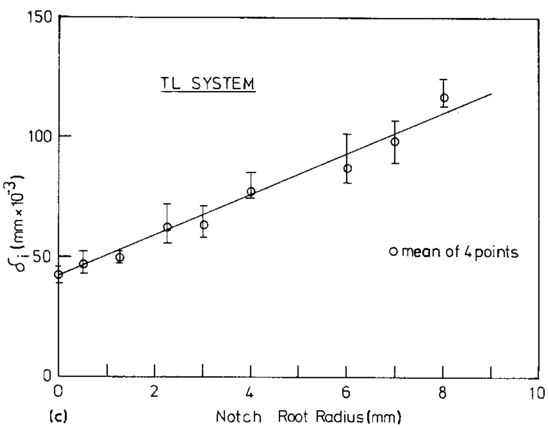


Figure 21 COD as a fracture criterion in wood: (a) calculation of crack tip opening displacement, δ from clip gauge measured values, V_g (after [42]); (b) and (c) for, respectively, LT and TL systems, show the variation of δ at fracture initiation, δ_i, with notch root radius.

that failure from a hole in a wooden member may not initiate at the point of maximum tensile stress (i.e. at the apex) but will rather fail due to a combination of moderate tensile stress and high shear stress which is achieved in the LT system at a point approximately 20° around the hole circumference from the apex (see [9] for details). If failure does occur at this point then, for large hole diameters, δ_i is no longer fully representative of the material at the notch tip, and a limiting value of δ_i would be expected.

A linear relationship between the crack tip opening displacement at failure initiation, δ_i, and the notch root radius, r, is an interesting observation, and one that suggests a strain-controlled fracture criterion, according to a method described by Knott [42]. Thus one can, in simplistic terms, imagine the element at the notch tip as a miniature tensile specimen of gauge length $2r$. At fracture initiation, this element will have reached its breaking strain, ϵ_f, which is given by $\epsilon_f = \delta_i/2r$. The linear plots shown in Fig. 21 therefore indicate a constant value of $\delta_i/2r$, and hence a constant breaking strain.

For both the TL and LT systems a fracture strain of approximately 0.5% is indicated, which is lower than the values of around 1% obtained from small, thin specimens by Ifju [23], Kellog and Ifju [22], and very recently by Ansell and Harris [44]. A reason for this discrepancy may be that the curvature of the notch leads to an overestimate of the gauge length. Steida [45] has shown that there is no significant difference between the fracture stresses of specimens having round ended notches and slots of equal width and length, but the imaginary tensile specimen is, in our case, necessarily thinner in the centre (i.e. at the apex of the notch) and this may cause a certain strain concentration at this point. Thus, the extension indicated by δ_i is not accomodated uniformly along the entire gauge length $2r$ and a lower value of failure strain could result. Alternatively (or additionally), the difference between the breaking strains obtained here and those recorded by other authors may be a function of differences in the stress state. Thus, the thin specimens used in other investigations [22, 23] would be in a state of plane stress, whereas the centres of the thicker specimens used in this study are presumably in a state of plane strain – certainly the clip gauge will respond to events occurring over the central part of the crack front which the crack bowing studies (Fig. 12) suggest take place first. It therefore seems reasonable to speculate that the fracture strain of wood in plane stress may be higher than that in plane strain.

7. Conclusions

Experimental investigations on the fracture behaviour of pre-notched specimens of South African pine, tested at three crack plane orientations under both static and dynamic loading, have led to the following conclusions.

1. Cell size has a dominant influence on fracture mode for both static and fatigue conditions. Small-celled "latewood" shows superior toughness and fatigue resistance to large-celled "earlywood", despite failing by a predominantly intercellular failure mechanism.

2. A reliable ink-staining technique has confirmed unequivocally the existence of crack front bowing. This is associated with a higher measured elastic strain distribution from the crack tip under conditions of stress triaxiality, i.e. at the centre of thick specimens. Further, thinner sections in all the orientations studied show superior toughness and fatigue resistance.

3. SEM observations show the existance of irreversible cell deformation mechanisms, involving debonding, axial twisting and buckling, in thin specimens and close to the edge of thicker ones which are not in evidence in the central regions of thick specimens. These observations can be explained in terms of stress state variations with specimen thickness.

4. Crack opening displacement (COD) measurements, made at the point of crack initiation, vary linearly with notch root radius. A simplified analysis indicates this result to be consistent with a strain-controlled criterion for fracture. Measured fracture strains are comparable with other results reported in the literature when account is taken of the pre-notched, bulk wood situation.

Acknowledgements

The authors are grateful to the South African Council for Scientific and Industrial Research and the University of Cape Town for financial support; one of the authors (SWJB) wishes to thank the Trustees of the J. W. Jagger Foundation for a postgraduate scholarship. Thanks are also due to Professors D. H. Avery and H. Vermaas for stimulating discussions leading to the work, and to Drs J. D. Barratt, J. M. Dinwoodie and G. Jeronomidis and Professor H. Vermaas for constructive comments on the manuscript.

References

1. J. M. DINWOODIE, *J. Microsc.* **104** (1975) 3.
2. D. ATTACK, W. D. MAY, E. L. MORRIS and R. N. SPROULE, *Tappi* 8 (1961) 555.
3. A. P. SCHNIEWIND and R. A. POZNIAK, *Eng. Fract. Mech.* **2** (1971) 223.
4. J. M. DINWOODIE, *Composites* **2** (1971) 170.
5. E. J. GIBSON, Proceedings of the Conference on Civil Engineering Materials, Southampton, April 1969 (Wiley Interscience, London, 1971) pp. 427–41.
6. K. BORGIN, *New Sci.* **64** (1974) 556.
7. A. P. SCHNIEWIND and J. C. CENTENO, *Wood Fibre* 5 (1973) 152.
8. S. W. J. BOATRIGHT, MSc thesis, University of Cape Town, South Africa (1977).
9. S. W. J. BOATRIGHT and G. G. GARRETT, *Holzforschung* 33 (1979) 68.
10. BS 5447, "Methods of test for plane strain fracture toughness (K_{Ic}) of metallic materials" (British Standards Institution, 1976).
11. M. TOMIN, *Wood Sci.* 5 (1972) 118.
12. H. BAUMANN, in "Principles of Wood Science and Technology I", edited by F. P. Kollman and W. A. Coté (Springer-Verlag, New York, 1975).
13. G. A. COOPER, *Rev. Phys. Technol.* 2 (1971) 49.
14. G. R. IRWIN, *Trans. Amer. Soc. Mech. Eng. J. Appl. Mech.* **24** (1957) 361.
15. A. P. SCHNIEWIND, *Wood Fibre* 9 (1977) 216.
16. J. D. BARRATT and R. O. FOSCHI, *Eng. Fract. Mech.* 9 (1977) 371.
17. N. L. HARRISON, *Fibre Sci. Technol.* 6 (1973) 25.
18. I. FURUKAWA, H. SAIKI and H. HARADA, *J. Electron Microsc.* **23** (1974) 89.
19. G. R. DEBAISE, A. W. PORTER and R. E. PENTONEY, *Mater. Res. Stand.* 6 (1966) 493.
20. Z. KORAN, *Tappi* **50** (1967) 60.
21. G. R. DEBAISE, *J. Mater. JMLSA* 7 (1972) 568.
22. R. M. KELLOG and G. IFJU, *Forest Prod. J.* **12** (1962) 403.
23. G. IFJU, *ibid.* **14** (1964) 366.
24. B. A. JAYNE, *ibid.* **10** (1960) 316.
25. M. F. KANNINEN, E. F. RYBICKI and H. F. BRINSON, *Composites* 8 (1977) 17.
26. P. D. EWING and J. G. WILLIAMS, *J. Mater. Sci.* **14** (1979) 2959.
27. J. D. BARRATT, *Wood Fibre* 6 (1974) 126.
28. S. W. J. BOATRIGHT and G. G. GARRETT, *Eng. Fract. Mech.* **13** (1980) 107.
29. D. H. PAGE, F. EL-HOSSEINY and K. WINKLER, *Nature* **229** (1971) 252.
30. G. JERONOMIDIS, Proceedings of the 3rd International Conference on Mechanical Behaviour of Materials, Vol. 3, Cambridge, UK, August 1979, edited by K. J. Miller and R. F. Smith (Pergamon Press, Oxford, 1980) pp. 329–40.
31. J. E. GORDON and G. JERONOMIDIS, *Nature* **252** (1974) 116.
32. *Idem, Phil. Trans. Roy. Soc. London* A294 (1980) 545.
33. W. A. COTÉ (ed), Proceedings of the Advanced Science Seminar, New York, September 1964 (Syracuse University Press, New York, 1965).
34. A. P. SCHNIEWIND and J. D. BARRATT, *Wood Fibre* **1** (1969) 205.

35. R. C. TANG, *ibid.* 4 (1972) 210.
36. J. D. BARRATT and A. P. SCHNIEWIND, *ibid.* 5 (1973) 215.
37. J. L. LEGGATT, *Amer. Rail. Eng. Assoc. Bull.* 55 (1953) 161.
38. A. C. SEKHAR and N. K. SHUKLA, *Holz Roh- und Werkst.* 23 (1965) 434.
39. A. C. HORNER, W. C. LEWIS, E. J. RUBIE and L. W. WOOD, *J. Struct. Div. Proc. Amer. Soc. Civil Eng.* (September 1957) Paper no. 1361.
40. M. P. BROKAN and G. W. FOSTER, "Effect of rapid loading and duration of stress on the strength properties of wood tested in compression and flexure", Forest Products Laboratory, Madison, Report no. 1518 (1964).
41. R. W. PERKINS, *Forest Prod. J.* 17 (1967) 57.
42. J. F. KNOTT, "Fundamentals of fracture mechanics" (Butterworths, London, 1973).
43. A. E. GREEN and G. I. TAYLOR, *Proc. Roy. Soc.* A184 (1940) 181.
44. M. P. ANSELL and B. HARRIS, Proceedings of the 3rd International Conference on Mechanical Behaviour of Materials. Vol. 3, Cambridge, UK, August 1979, edited by K. J. Miller and R. F. Smith (Pergamon Press, Oxford, 1980) pp. 309–18.
45. C. K. A. STEIDA, *J. Mater.* 1 (1966) 560.

Received 22 January
and accepted 20 December 1982