

EXPERIMENTAL INVESTIGATIONS OF SIZE EFFECT ON FRACTURE TOUGHNESS OF METALLIC FOILS

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Abstract

The effect of thickness on the fracture toughness J_C of metallic foil was investigated experimentally. Double-edge cracked specimens, made of copper foils with thicknesses t ranging from 0.02mm to 1mm, were loaded in tension till fracture. The digital speckle correlation method (DSCM) was used to evaluate the strain fields around the crack tip, allowing determination of the J integral. The fracture toughness defined as the value of the J integral at cracking initiation was shown to first increase with increasing thickness, then to reach a maximum for a thickness of about 0.3mm and finally to decrease at larger thicknesses. Optical and scanning electron microscopy (SEM) results showed that this significant effect of thickness is essentially attributable to the change in the work required per unit area during crack tip necking and in the work required per unit area during material separation.

Keywords: Fracture Toughness; Thickness Effect; Microstructure; Metallic Foil; Digital Speckle Correlation Method.

1. Introduction

With the rapid development of modern technology, foil materials have found applications in the areas of materials and electrical industries such as microelectro-mechanical systems (MEMS) and integrated circuits. The mechanical behavior of such materials may be different from that of bulk materials due to size effects. Therefore, models and conclusions appropriate for bulk materials may not be applicable when analyzing foil materials. Certain essential problems and phenomena occurring during the fracture process of foils must be explored by experiments and numerical simulation methods in order to provide a better understanding of fracture mechanisms, especially their difference from bulk materials. For example, the effect of the thickness of the foils on the fracture behavior remains an open issue.

During last two decades, several investigations of the mechanical properties of foils have been carried out both experimentally and theoretically. For example, Sutton and Han et al. [1,2] experimentally studied stationary crack-tip deformation fields, evaluated the J integral, and investigated the three-dimensional effects near a crack tip in thin sheet using computer vision. Studies [3,4] have demonstrated that size effect on the material properties of foil materials was attributed to dimensional and microstructural constraints. Further research concerning size effect on fatigue behavior of thin films has shown that the mechanical behavior was closely related to the thickness and grain size of the films [5-9]. Choi et al. [10] analyzed size effect on the mechanical properties of thin polycrystalline metal films, and developed a model that correctly predicts the observed influence of film thickness and grain size on stress evolution during thermal excursions. Pardoen et al. [11,12] investigated size effect on the fracture toughness of aluminum thin plates of 1-6mm thickness from tensile testing of double edge notched tension (DENT) specimens. Their research showed that thickness indeed influences fracture toughness. Kang et al. [17,18] measured and calculated the fracture toughness of copper foil by an improved digital speckle correlation method (DSCM).

From the above review, we note that most of the existing work has focused on mechanical size effects in thin films whose thickness was usually greater than 1mm or less than 1mm but supported by a substrate. Nowadays, many modern electronic devices or packages involve very thin “free-standing” films (usually much thinner than 1mm) for which the mechanical and fracture behaviors are critical in analyzing the performance of whole structure. However,

not enough is known about the complex size effect in such foils that is required for optimal design and manufacture. The present work aims at addressing this problem. The DSCM as previously established [1,2,13-18] is used to obtain the displacement and strain fields near the crack tip region, allowing determination of the fracture toughness, characterized by J_C and defined as the value of the J integral at cracking initiation, of specimens with thicknesses ranging from 20 μm to 1mm. Then macroscopic and microscopic examinations by optical and scanning electron microscopy (SEM) are used to explore the fracture profile and the specimen surface of the copper foils. The influence of thickness effect and microstructure on the fracture toughness of the foil is discussed in detail.

2. Experimental procedure

The specimens employed in this study were made of T2 copper foil with thicknesses ranging from 20 μm to 1mm. All were produced by the same processing technology, and thus had the same chemical composition and content. The Young's modulus E and Poisson's ratio μ of the material were about 108.5GPa and 0.334, respectively. The thin foils were processed with different thicknesses and double-edge cracks. The crack on each specimen was made as follows: a line-incisor with a radius of 0.1mm was used to make an initiatory crack and then a sharp razor was used to make the crack tip. The radius of the razor was approximately 25 μm . The validity of this pre-cracking method has been justified by Pardoen et al. [11]. The dimensions are shown in Fig. 1. Additionally, the specimens were classified into H-specimens and V-specimens, according to the relationship between the rolling direction and the crack path. H-specimens were those with a crack path parallel to the rolling direction; V-specimens were those with a crack path perpendicular to the rolling direction.

The DSCM (also known as digital image correlation method) was used in this work for measuring the fracture toughness J_C of the specimens. The process in DSCM includes recording, digitizing and processing two speckle patterns (or images) of an object in two different deformation states to yield in-plane displacement components and in-plane displacement gradients. The two speckle images, known as reference and deformed speckle patterns, of the surface of the object are usually captured before and after loading. A small speckle area in the undeformed speckle pattern is taken as a reference subset and the one corresponding to the reference subset in the deformed speckle pattern is defined as the target subset. Differences between the two subsets include translation and distortion information about the object. Finally, the J integral can be calculated based on the strain field in the region around the crack tip, using

$$\begin{aligned}
 J = & \frac{1}{2} \int_{\Gamma} \left[\frac{E}{1-\mu^2} \left(\frac{\partial u}{\partial x} + \mu \frac{\partial v}{\partial y} \right) \frac{\partial u}{\partial x} + \frac{E}{1-\mu^2} \left(\frac{\partial v}{\partial y} + \mu \frac{\partial u}{\partial x} \right) \frac{\partial v}{\partial y} + G \left(\frac{\partial u}{\partial y} + \frac{\partial v}{\partial x} \right)^2 \right] dy \\
 & - \int_{\Gamma} \left[\frac{E}{1-\mu^2} \left(\frac{\partial u}{\partial x} + \mu \frac{\partial v}{\partial y} \right) \frac{\partial u}{\partial x} + G \left(\frac{\partial u}{\partial y} + \frac{\partial v}{\partial x} \right) \frac{\partial v}{\partial x} \right] dy \\
 & + \int_{\Gamma} \left[\frac{E}{1-\mu^2} \left(\frac{\partial v}{\partial y} + \mu \frac{\partial u}{\partial x} \right) \frac{\partial v}{\partial y} + G \left(\frac{\partial u}{\partial y} + \frac{\partial v}{\partial x} \right) \frac{\partial u}{\partial x} \right] dx
 \end{aligned} \tag{1}$$

where Γ is an anticlockwise curve that surrounds the crack-tip, u and v are the components of the displacement vector in the x - and y - direction, respectively. Relevant derivation of J_C can be found in [18] and we omit those details here for conciseness.

A double charge coupled device (CCD) system equipped with zooms (Fig. 2) was designed to accurately determine the value of J_C , i.e., J integral when the crack tip initiates. One CCD system was used to observe the crack tip in real time and another to serially capture the digital speckle pattern. The pattern captured just before crack tip initiation was taken as the deformed image. Several methods of detecting initiation of fracture have been described in the literature. Here we refer to the method given by E-Isoudani [19], namely thumbnails technique, because the precision of this technique is high enough for our experimental purpose and can be used in the experimental research step by step. Moreover, it has been

used by many other researchers. The experimental method is schematically summarized in Fig. 3.

The experiment in this work also includes metallographic examination of the specimen surface and fracture profile by optical microscopy and SEM. In particular, the surface and fracture profile of the specimens, including the amount of roll marks, the degree of crack-tip necking and the macro-appearance, were observed by optical microscopy before and after deformation. The micro-structural features of the fracture profile were examined by SEM (Philips, XL30 ESEM). These tests were performed at room temperature.

3. Experimental results and discussion

3.1 Toughness/thickness curve

Based on DSCM experimental results, the fracture toughness J_C of the specimens is presented in Fig. 6 as a function of specimen thickness, showing that J_C strongly depends on thickness. In a so-called stage I (see Fig. 4), J_C is shown to increase with an increase in thickness, then to reach a maximum at the thickness $t \approx 0.3\text{mm}$. Conversely, J_C decreases as the thickness increases within the so-called stage II (see Fig. 4). It is also evident that the J_C values of the V-specimens are always greater than those of the H-specimens at any given thickness. Moreover, it should be noted that Fig. 6 is similar to the plots presented in many textbooks such as that of Kanninen and Popelar [20].

3.2 Analysis and discussion of J_C versus thickness

The experimental results from macroscopic and microscopic examination of the fracture profile show that the degree of necking at the crack tip and the micro-fracture appearance in front of the crack tip (Fig.4) are different for different thicknesses. The degree of crack-tip necking is related to the work required per unit area during crack-tip necking. The micro-fracture appearance is linked to the work required per unit area during material separation. The value of the fracture toughness is the sum of the two contributions. Additionally, it is known that the work per unit area for crack-tip necking is greater at a larger degree of crack-tip necking and the work per unit area for material separation increases when the amount of slip bands and tear ridges increases. Moreover, the work per unit area for material separation in the case of ductile fracture is greater than that in the case of brittle fracture. Therefore, what follows is concerned with the relationship between the fracture toughness J_C and the abovementioned fracture features, dividing the curve into two stages (the stage during which J_C values increase with the increase in thickness is denoted as stage I, the remaining is denoted as stage II).

In stage I, the results of macroscopic examination show that the degree of necking at the crack tip increases with increasing thickness in the macroscale. SEM micro-fractographs show that specimens with the smallest thickness show that brittleness occurs with shear mode fracture feature. The specimen with a thickness of 0.1mm shows shear mode as well as dimple-ductile fracture features, with very shallow dimples elongated in the direction pointing to the surface. Specimens with thickness ranging from 0.2mm to 0.3mm show ductile fracture features with slip bands and lenticular dimples. Additionally, in the case of ductile fracture, as the thickness increases, the amount of slip bands, tear ridges and the size of the dimples all increase, and the dimples are become deeper. This indicates that the fracture mechanism changes from a mixed brittle/shear-mode fracture to a mixed shear-mode/ductile fracture, and then to a mixed ductile fracture with slip bands, tear ridges and lenticular dimples. It is considered that the changes in the crack-tip necking degree, the dimple diameter and depth, the amount of slip bands and tear ridges, and the fracture mechanism are mainly due to the influence of a reduction in micro-defects (i.e. roll marks on the surface and defects caused by rolling) and an increase in the plastic deformability of the material itself (the influence of the stress triaxiality is assumed to be negligible when thickness is very small). The thinnest specimen show brittleness occurring with a shear mode fracture pattern with no necking degree, because their deformation potential is the lowest while the influence of the micro-defects is the strongest. It thus exhibits the lowest fracture toughness due to its toughness being affected by the work involved for material separation only, which is the least in comparison with specimens of greater thickness. Then as the thickness increases, the influence of the micro-defects reduces but the deformability of the specimen

increases. This results in an increase of the crack-tip necking degree and a change in the fracture mechanism. These fracture features indicate that the work required per unit area for crack-tip necking and the work per unit area for material separation both increase with increasing thickness. Consequently, fracture toughness, which is comprised of the above two contributions, increases with increasing thickness in stage I.

In stage II, however, the degree of crack-tip necking decreases with increasing thickness, and the fracture mechanism changes from mixed ductile fracture with slip bands, tear ridges, and lenticular dimples to mixed ductile fracture with tear ridges and well-developed equiaxed dimples only. Moreover, the fractographs show that the dimple diameter and depth become respectively larger and deeper as the thickness increases. However, the amount of slip bands and tear ridges decreases with an increase in thickness. In addition, an increase in the stress triaxiality may lead to a decrease in the crack-tip necking degree and the amount of slip bands, as well as a change in the fracture mechanism under consideration. In this case, an increase in the stress triaxiality will lead to a decrease in the degree of necking and an increase of the deformation resistance when crack extension begins. The increase of the dimple diameter and depth can be attributed to the increase in the plastic deformability of the material and the fact that void growth is favored by higher stress triaxiality, which may lead to the decrease in the amount of tear ridges. These fracture features indicate that an increase in thickness will lead to a decrease in both the work per unit area for crack tip necking and the work per unit area for material separation. Therefore, fracture toughness decreases with an increase in thickness during stage II.

3.3 Comparison between V-specimens and H-specimens

The results of macroscopic examination of the surface showed evidence of roll marks along the rolling direction, and the grains are also elongated along the rolling direction. These two factors are thought to reduce the fracture resistance and to be responsible for the fact that the J_C values of the V-specimens are always greater than those of the H-specimens when the thickness is the same. Firstly, the presence of the roll marks indicates that the micro-defects occur along the rolling direction. In the case of the H-specimens, the direction of the roll marks is perpendicular to that of the tensile load and parallel to that of the main crack, so it is easier for micro-defects to induce coalescence of the inner or surface micro-cracks to form “macro-cracks” when subjected to tensile load. Furthermore, the grain flow direction is parallel to the main crack in the H-specimens. As a result, a crack in H-specimens is easier to initiate because the fracture plane follows the plane of lowest crack resistance, with less grain boundaries to cross.

4. Conclusions

A comprehensive experimental study of the effect of thickness on the fracture toughness of metallic foil has been carried out. Fracture toughness, characterized by J_C , of copper foils with thicknesses from 0.02 mm to 1 mm was evaluated using DSCM. Experimental results indicate that the value of J_C depends strongly on thickness. That value is shown to increase with increasing thickness in a so-called stage I, then to reach a maximum at the thickness $t \approx 0.3$ mm and finally to decrease with an increase in thickness within the so-called stage II. This significant effect of thickness is essentially attributed to the change in the work required per unit area during crack-tip necking and in the work required per unit area during material separation. The major factors affecting these two contributions are considered to be related to the combined action of the microstructure, the presence of inner micro-defects and roll marks, and the level of stress triaxiality. In the present work, the significant effect of thickness is specifically associated at low thickness with (i) a change in failure mode, from a more brittle mechanism to a more ductile mechanism involving void growth, and (ii) a combined evolution of the degree of crack tip necking. At greater thicknesses, variation of stress triaxiality is thought to be the major factor responsible for the decrease in fracture toughness with thickness.

Acknowledgements

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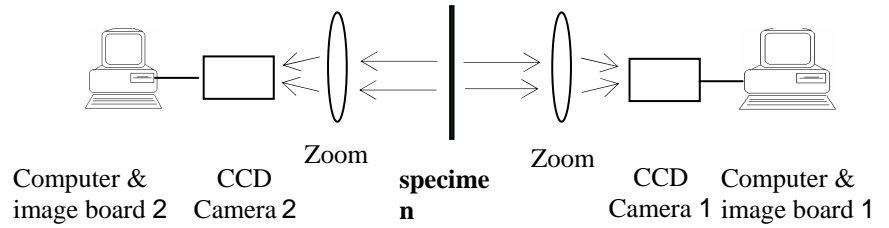
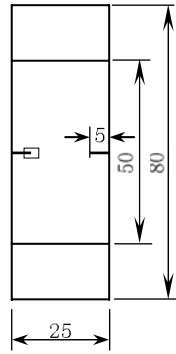


FIGURE 1. Dimensions of specimen and

FIGURE 2. Double CCD system

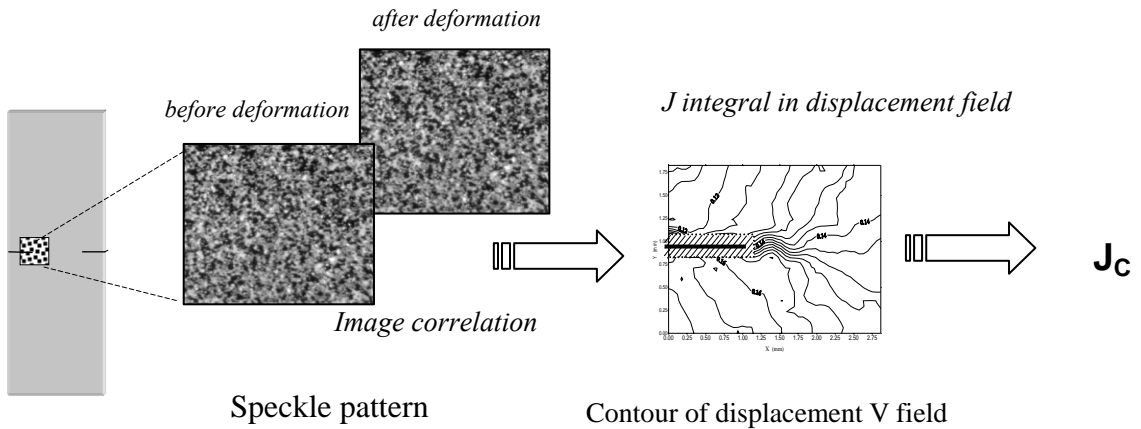


FIGURE 3. Schematic illustration of J_c evaluation processing

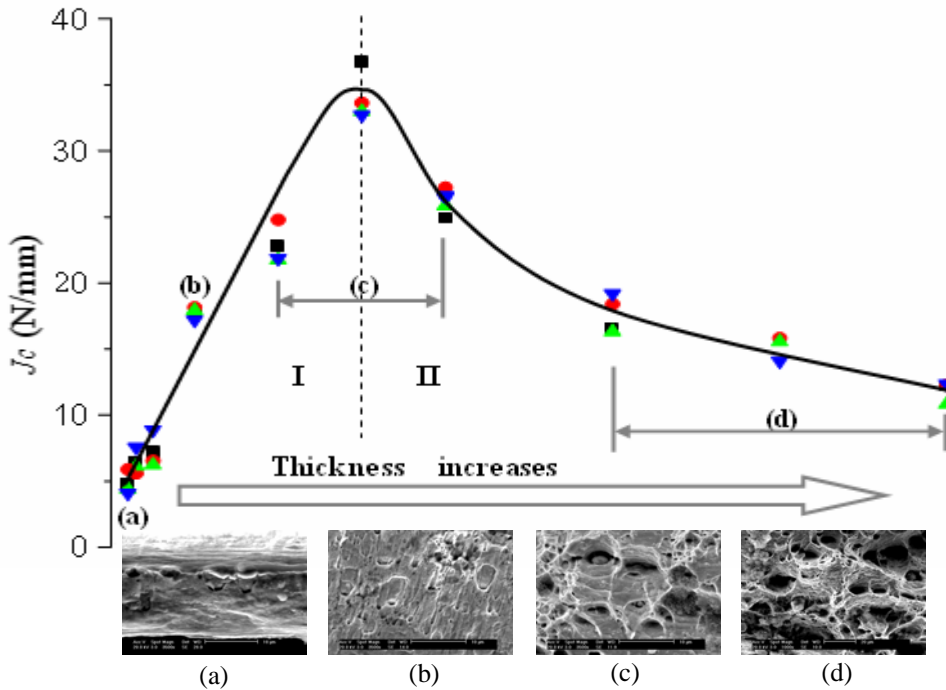


FIGURE 4. Relation between J_c and thickness as well as fracture mechanism. (a) Brittle fracture coexists with shear mode fracture; (b) shear mode fracture coexists with ductile fracture with shallow dimples; (c) Mixed ductile fracture with slip bands, tear ridges and lenticular dimples; (d) Mixed ductile fracture with tear ridges and well-developed equiaxed dimples. It shows that the fracture toughness is associated with the fracture mechanism.