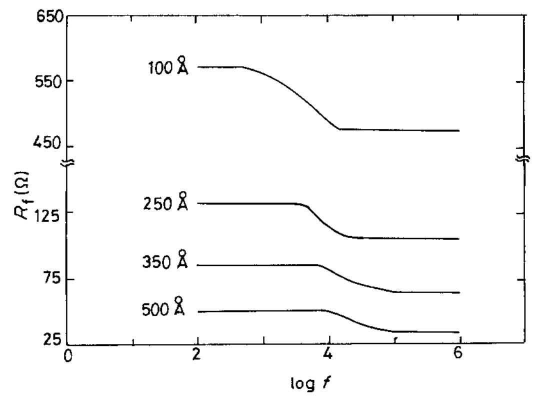

Figure 2 Variation of samarium film resistance (R_f) with frequency (f) for film thicknesses, 100, 250, 350 and 500 Å.

response is flat at lower frequencies (< 10 kHz) but exhibits a sharp decrease in resistance with an increase in frequency and finally attains a constant minimum value. Similar behaviour has also been observed for PbS films [15]. At lower thicknesses, metallic films lack electrical continuity between the islands and the inter-island space is filled with either air or some dielectric material. Hence, this metallic structure in which the islands are separated by small spaces can be treated as equivalent to a series of capacitors [16] and hence might lead to the frequency dependence of the resistance.

The study of high-frequency response characteristics on the a.c. resistance of thin metal films may confirm the validity of this simple model [16], which will be reported in a future communication.

Acknowledgments
One of the authors (PVA) is grateful to the Council of Scientific and Industrial Research, for the award of a Senior Research Fellowship. The financial assistance from the University Grants Commission, New Delhi, is gratefully acknowledged.

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*Received 27 April
and accepted 20 May 1981*

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Comments on the layer removal method for measurements of residual stresses in plastics

The phenomenon of residual stresses and their significance in polymeric materials have only recently received scientific and technological attention. The measurement of residual stresses in polymeric materials is still a subject for investigation. There are several methods reported for the estimation of residual stresses. They are based either on the disturbance of the state of stress equilibrium or on stress sensitive physical properties. The former include such methods as environmental stress-cracking [1], hole drilling [2], split cutting [3] and layer removal [4]. Other methods include surface hardness [5], stress relaxation [6] and bi-refringence [7]. None of these methods

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is capable of describing the complete state of residual stresses. However, the "layer removal" method seems to be the most suitable one for polymeric materials and it has recently become the most common method to be used (e.g. [8–13]). This method enables the estimation of residual stresses distribution and as such is not restricted to the surface stresses. According to this method, stresses, as a function of distance from the specimen surface, are measured by successive removal of uniform layers from one surface of a rectangular bar. The resulting curvatures permit the calculation of the residual stresses. Treuting and Read [8] have worked out the relation between residual stresses and the measured curvatures. Assuming no directional effect in the plane of the specimen, the residual stresses can be described as follows:

$$\sigma_x(Z) = \sigma_y(Z)$$

$$= \frac{-E}{6(1-\nu)}\left[(Z_0+Z_1)^2 \frac{\mathrm{d}\rho_x(Z_1)}{\mathrm{d}Z_1} + 4(Z_0+Z_1)\rho_x(Z_1) - 2\int_{Z_1}^{Z_0} \rho_x(Z)\,\mathrm{d}Z\right], \quad (1)$$

where E is the elastic modulus, ν is Poisson's ratio, ρ_x is the curvature parallel to the x-direction, x and y are the longitudinal and traverse directions of the plate, $Z = \pm Z_0$ are the initial upper and lower surfaces of the specimen and $Z = Z_1$ is the new upper surface after each layer removal. Treuting and Read [8] have assumed in their derivation that the elastic mechanical properties, E and ν, are independent of Z_1, namely, constant throughout the specimen thickness. The purpose of the present discussion is to question this assumption and hence the accuracy of this method for determining residual stresses in polymeric materials in general and further to discuss its applicability to the analysis of end products.

In the present work, flat plates (11 cm x 11 cm x 0.5 cm) of Noryl-SE1 (General Electric modified polyphenylene oxide) were annealed at 145° C (above its glass transition temperature, T_g) and quenched into ice water. The residual stresses were measured in rectangular 6 cm x 1 cm x 0.5 cm bars which were cut from the quenched plates. The calculated profile, based on the method described above, is presented in Fig. 1. It is characterized by a

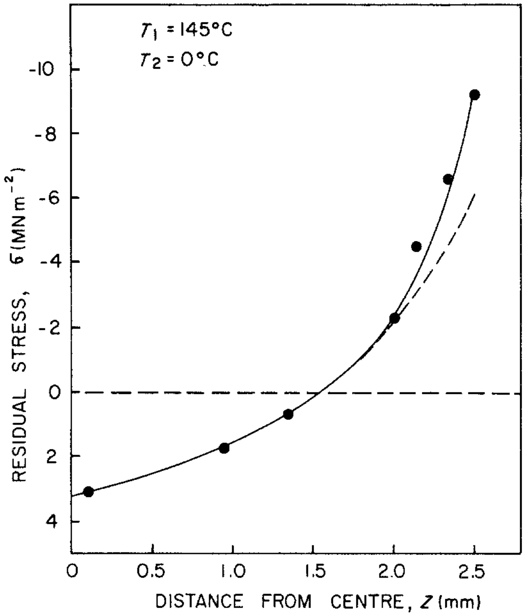

Figure 1 Residual stresses distribution through the thickness of Noryl plate annealed at 145° C and quenched into ice water. (a) Based on constant E (solid line), (b) based on variable E (dashed line).

steep gradient of compressive stresses at the surface layers and tensile stresses in the interior. It is, in general, a typical profile expected in flat plates which were rapidly cooled through the glass transition temperature of the material.

Using the same method of layer removal enables one to measure the profile of mechanical properties throughout the thickness of the specimen. The tensile mechanical properties (tensile modulus, E, stress at break, σ_B, strain at break, ϵ_B) of the quenched Noryl were found to gradually increase with increasing distance from the surface. Concentrating on the changes in the elastic modulus (see Fig. 2), it is interesting to note that its value is always larger compared with that of annealed and slowly-cooled specimens, which are free of residual stresses. The modulus increases with distance from the surface and levels off at about 70% of the specimen thickness. These measured moduli are accumulative values, characteristic of specimens after the removal of progressively thicker surface layers. However, when the local values are estimated, by applying the "rule of mixtures", values as low as 1.1×10^3 MN m^{-2} at the surface and as high as 1.5×10^3 MN m^{-2} in the interior are obtained. These differences are

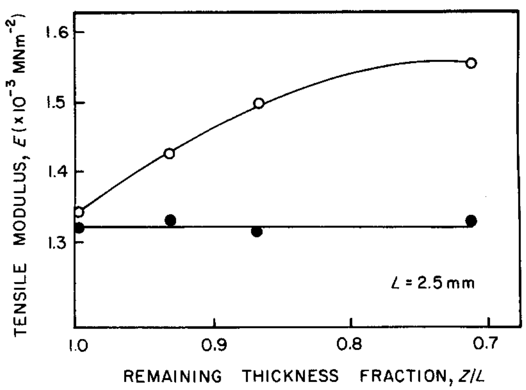

Figure 2 Accumulative tensile modulus profile in Noryl plate annealed at 145° C followed by (a) quenching into ice water (○), (b) slowly cooled to ambient temperature, (●).

much larger than those quoted in the literature for quenched and annealed materials. However, it should be noted that the latter are always averaged over the whole cross-section of the specimen. These large variations in the modulus are in accordance with the profile observed for the density of such specimens [14].

It has been demonstrated that the modulus of elasticity is not constant throughout the thickness of quenched samples, in which a profile of residual stresses has also been developed. In the Treuting and Read [8] derivation the residual stresses, $\sigma_x(Z_1)$, are linearly proportional to the modulus and thus any changes in the latter will directly influence the calculated values of residual stresses. Since the modulus was observed to change mainly in the surface layers, where the largest changes were also observed in the residual stresses, large inaccuracies are introduced in these calculations by assuming a constant modulus. When the modulus profile presented above is taken into account, the calculated residual stresses will change accordingly and their resulting profile will not be as steep in the outer layers (see dashed line in Fig. 1).

The treatment of data based on a single value of the modulus is even more questionable when residual stresses are being measured in end products such as injection-moulded or extruded material. These products are inhomogeneous and anisotropic in their properties, as a result of such factors as variation in thermal history and molecular orientation. In such products the local elastic modulus varies through the thickness as well as across the product (effected by machine direction, distance from gate, etc.). For example, in oriented sections the modulus is actually much higher than the average value commonly used for calculations; hence, the residual stresses in these sections are much higher than the calculated ones.

In conclusion, to avoid misleading analyses, the Treuting and Read [8] derivation, in its present form, for calculating residual stresses in polymeric materials should be used very cautiously. Therefore, when a three-dimensional profile of residual stresses is looked for, the three-dimensional distribution of elastic mechanical properties should be accounted for.

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Received 27 April
and accepted 20 May 1981

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