Young's modulus of cast HMS 2112- C_f composite—prediction and ultrasonic evaluation

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Abstract. Various models for the prediction of strengthening mechanism of metal matrix composites (MMCs) containing either fibres or particulates are analysed. Assuming that the matrix strengthening by dislocations could be treated as equivalent to the effect of different volume fraction of dispersoids, as well as by considering the effect of morphology of reinforcement on the Young's modulus, an expression for Young's modulus for MMCs has been derived. The Young's modulus values thus predicted, using this model, have been validated by ultrasonically-derived values of Young's modulus of an Al-alloy matrix composite containing 5, 8 and 12 wt% chopped carbon fibre (C_f) dispersoids, in as cast and extruded conditions. Further, the theoretically- and ultrasonically-derived Young's modulus of cast Al-alloy- C_f composites with 5 and 8 wt% C_f have been found to be comparable with the reported values of Young's modulus for these weight fractions.

Keywords. Strengthening mechanism; Young's modulus; metal matrix composites; ultrasonic testing; weakening factor.

1. Introduction

Strengthening mechanism in metal matrix composites (MMCs) has been found to be different from those in monolithic alloys (Taya and Arsenault 1989). It has been a subject matter of intense research, to understand the mechanical behaviour of MMCs (Arsenault 1990; Lloyd 1994; Kwon and Dharan 1995) for continuous fibre composites as well as for discontinuous dispersoids (Mc Denals 1995). These strengthening mechanisms have been classified under five basic models starting from simple rules of mixtures (ROM), originally proposed by Voigt, to more rigorous models such as Eshelby's and variational principle models. Since elasticity is sensitive to volume fraction of dispersoids, most of the models have predicted elastic modulus of composites and validated it through experiments. In addition, shape, size and interdispersoid spacing as well as continuity of phase also dictates the Young's modulus (Cox 1952; Arsenault and Liu 1987; Taya and Arsenault 1987; Arsenault 1990). All the theoretical models have used one or two of these parameters to predict the modulus of the composite depending on the nature of dispersoid.

Comparing experimental values, the ROM model predicts reasonably well the longitudinal properties of

composites containing unidirectional continuous reinforcements. But transverse properties of such composites or those of short fibre containing composites cannot be explained on the basis of this model.

The shear lag model originally proposed (Cox 1952) assumed that (i) short fibres of uniform aspect ratio are uniformly distributed in the material, (ii) they are aligned in the direction of loading and (iii) shear stress is proportional to the difference in the axial displacement; the applied strain being not the same as that considered in the ROM model. Due to this inconsistency between the two models, both models are used together in view of their simplicity. But shear lag model gives poorer approximation for short fibres of smaller aspect ratio. Eshelby's model (Arsenault and Liu 1987) considered inclusions or inhomogeneity (e.g. dispersoid) of ellipsoidal shape to be embedded in an infinite elastic domain. A mathematical solution was obtained for the system containing single inclusion inhomogeneity and inhomogeneous inclusions without the interaction of inhomogeneities (dispersoids) and subjected to uniform elastic strain. However, in reality, volume fractions of inhomogeneities in composite are finite. Hence the interactions between inhomogeneities become important. This interaction was taken into account by including a term and then an approximate solution for the stress field was obtained using inhomogeneous inclusion problem. Thus the inhomogeneity problem was reduced to an inclusion

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called 'equivalent inclusion problem'. Analysis carried out on the basis of the above, yielded an approximate formula for the effective elastic modulus of a composite. This rigorous model gave better prediction compared to the other two models.

On the other hand, for particulate dispersoids, Tsai and Halpin (Lloyd 1994) modified the ROM by incorporating suitable terms for geometry, distribution and loading condition. The equations derived by them are valid for low volume fraction of fibres and for discontinuous-dispersoids oriented in the loading direction. There are other macro mechanical models proposed (Mori and Tanaka 1973; Kwon and Dharan 1995) for determining the effective moduli of the composites. But it is reported that these macro mechanical approaches pose difficulty in obtaining good agreement with experimental data for the effective moduli of the composite when the volume fraction of the dispersoid is high. Recently, Kwon and Dharan (1995) have incorporated a concept of continuity of the second phase into micro mechanical model. This approach shows excellent agreement with observed elastic moduli values of tungsten carbide-cobalt (Wc/Co) system. These models are applicable mainly for particulates.

While the above models consider various parameters such as residual elastic stresses, amount, shape and size of dispersoids independently or in conjunction, there are some limitations in all the models discussed above such as they are applicable only for continuous fibres or particulates up to a certain volume fraction of the dispersoid. None of these models seem to take into effect dislocation density obtained in MMC due to the mismatch between the matrix and the dispersoids. Since the morphology of dispersoid along with the amount of the dispersoid dictate the strength properties of composites, it is possible that Young's modulus of composite also follows this. Comparison of the mostly used three-elastic modulus of Al-SiC system with the experimentally observed values clearly indicate that amount of dispersoid is the dominant factor for the increasing elastic modulus in the composite but is insensitive to the type of reinforcement (particulate or fibre). Hence, the objective of the present study is to bridge this gap, particularly in respect of effect of dispersoid morphology on the Young's modulus of composite along with the stress associated with dislocation densities, as well as to arrive at a unified simple model applicable for both continuous and discontinuous dispersoids without too much involved mathematical treatment. Further, the predicted values are validated by actual measurement of Young's modulus through ultrasonic testing of Al-alloy containing chopped carbon fibres processed by stir casting method (Pai *et al* 1994). These values are compared with the experimental values obtained by destructive testing (Satyanarayana *et al* 1994).

2. Present study

2.1 Theoretical aspect

A weakening effect factor is assumed, which arises due to the breaking of dispersoids, debonding and disorientation of dispersoids in the composites, which is compensated to a great extent by the presence of dislocation density around the particles. The dislocation density associated with the small volume fraction has similar effect as it would have been with the presence of higher volume fraction of dispersoid without any dislocation. Consequently, the effective volume fraction of dispersoid would be V_f' instead of V_f , and the expression for Young's modulus following ROM would be

$$E_c = E_m(1 - V_f) + XE_fV_f,$$

where E_c , E_m and E_f are Young's modulus of composite, matrix and fibre, respectively and X is the weakening factor given by $X = a \cdot \exp(b \cdot V_f)$, where, ' a ' is a factor indicating effectiveness resulted by the orientation of dispersoid (i.e. poor for particulate), good bonding at the interface (better for fibres), creep resistance (more for fibres) etc, and ' b ' is a characteristic constant and

Figure 1. Variation of 'X' with volume fraction (%).

function of l/d ratio (Vishnu Nampoothiri 1995). In order to arrive at a value for X , initially, experimental data of Al-20 V% SiCp, processed by powder metallurgy (Harrigan 1987), was used for deriving a mathematical expression for Young's modulus since not much data on short fibre was available. This gave a value of 0.48 for X in this composite. Then, different values of X for different V_f were calculated and a plot (figure 1) of X vs V_f was made assuming $X=1$ at $V_f=100$. It can be seen that initial value of X is linear at lower volume fraction followed by a steep increase at higher volume fraction, thereby showing an exponential relationship with volume fraction. The term $b \cdot V_f$ is indicative of dislocations and we then have

$$V_f' = V_f e(b \cdot V_f).$$

Further, b is assumed to increase with l/d up to a limiting value of l/d , say about 55. Beyond this, corresponding to a long line of dislocation, there could be a parallel free line through which matrix flow can occur under a load and hence increase in the value of ' b ' may be negligible. We also assume variation of b with l/d in a manner which is followed by Young's modulus in Tsai-Halpin equation (Halpin 1990). This equation thus demonstrates the initial importance of the aspect ratio as well as ineffectiveness of the same beyond a limit as depicted in figure 2.

Using the proportions taken from figure 1, ' b ' is given by

$$b = 0.9 + 0.02 \sqrt{23.04 - \{5 - 0.2(l/d)\}^2}$$

for $l \leq l/d \leq 25$,

and $b=1$ for $l/d > 25$. The value of a is obtained as 0.385 (Kwon and Dharan 1995) for particulates (discontinuous dispersoids), while for fibres it is assumed to be near unity on the lines of 'degree of continuity' defined by Kwon and Dharan (1995) for improved bonding which also helps in prevention of crack propagation through fibre orientation.

However, effectiveness of ' a ' for fibres remains near unity at moderate values of V_f say up to 50%. For further increase in V_f , interdispersoid distance becomes so small that the probability of their touching each other increases and crack propagation may become easier. At extreme points ($V_f=90\%$) ' a ' may be reduced to lower values.

2.2 Experimental

Extruded HMS 2112 (equivalent to 7075 Al alloy i.e. Al-Zn-Mg system) C_f composite containing 5, 8 and 12 wt% C_f of size 19 mm dia and 20 mm long were used for the ultrasonic evaluation. To eliminate the effect of surface roughness on the ultrasonic evaluation, the test

surfaces were machined to have good surface finish.

Krautkammer ultrasonic tester with longitudinal waves generated using a straight beam probe of 5 MHz frequencies and 5 mm crystal dia was used. The ultrasonic waves were sent through the composite in the direction of fibre alignment. The velocities were measured using the microprocessor, by the time of flight as an ultrasonic pulse through the specimen of known thickness. Thus, the velocity (v) was derived on the basis of the thickness of the specimen and the time of flight. Knowing the values of density (ρ) and Poisson's ratio (μ) of the composite, Young's modulus of the composite was obtained (Vishnu Nampoothiri 1995) using the equation (Krautkammer and Krautkammer 1987)

$$E = \frac{\rho(1-\mu)(1-2\mu)v_2}{1-\mu}.$$

3. Results and discussion

Figure 3 shows a comparison of experimental data with

Figure 2. a. Variation of Young's modulus of composite with aspect ratio of dispersoid and b. variation of ' b ' with aspect ratio of dispersoid.

Table 1. Young's modulus by rule of mixtures, present theory and ultrasonic evaluation.

Specimen number	Wt%	$V_f\%$	Moduli (Gpa)		Ultrasonic moduli (Gpa)	Available data by tension test (Gpa)
			Rule of mixtures	Present study		
1	5	7.714	81.1853	78.86	73.06	75–80
2	8	12.12	87.574	84.865	83.687	80–85
3	12	17.18	95.810	93.694	66.00	–

Figure 3. Young's modulus of a SiCp/7091 Al composite.

predicted values of Young's modulus for 7091 Al–SiCp composite (Jeong *et al* 1994). It can be seen that the prediction according to ROM for the composite is deviating too much from the experimental values while the same lie on the line based on present model.

With a view to check its validity for short fibres, Young's modulus of HMS 2112– C_f composite with 5, 8 and 12 wt% C_f were calculated using the present model with $E_f = 215$ Gpa, $E_m = 70$ Gpa, $l/d = 6.928$ and $a = 0.8$. The values of Young's modulus are given in table 1 together with those experimental values which were obtained through ultrasonic studies. Also the table contains YM values reported for these composites by tensile testing (Pai *et al* 1994; Satyanarayana *et al* 1994). It can be seen that there is an excellent agreement of predicted values of YM and ultrasonic moduli for the

Figure 4. Macro photograph of HMS 2112– C_f extruded sample showing surface defects.

specimens with 8 wt% C_f . The reduction in experimental value from the value of YM for 12 wt% C_f composite could be due to defects in the specimen as evident from the micrograph (figure 4). Further, there exists marginal difference between the theoretically and ultrasonically measured modulus for 5 wt% C_f composite, even though the values lie in the range obtained by tensile testing. This could be due to the presence of porosities in the ultrasonic test pieces, as evident from the testing.

4. Conclusions

A theoretical model has been proposed for strengthening mechanism in metal matrix composites (MMCs) taking into account the dislocation density obtained in MMCs due to the mismatch between the matrix and discontinuous dispersoid. Moreover, this model is applicable for both continuous dispersoids containing MMCs, unlike in earlier models. Further, predicted values are validated by the actual measurement of Young's modulus through ultrasonic testing of HMS 2114– C_f composites. These values compare well with the experimental values obtained by destructive testing of the MMCs.

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