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VALIDATION OF MODELS FOR THE PREDICTION OF THE THERMAL EXPANSION COEFFICIENTS FOR FIBRE REINFORCED TITANIUM COMPOSITES

by

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SUMMARY

This report is concerned with the validation of models that predict the thermal expansion coefficients of Ti-6Al-4V titanium alloy reinforced with Sigma silicon carbide fibre reinforced composites in unidirectional and laminated cross-ply forms. The validation is achieved by measuring the thermal expansion coefficients of the composites and by comparison with the predictions of the thermal expansion coefficients obtained from selected micro-mechanical models for unidirectional and cross-ply laminates. For unidirectional composites the axial and transverse thermal expansion coefficients are estimated in terms of fibre and matrix properties by making use of the well known concentric cylinder model. The temperature dependence of Young's modulus and thermal expansion of the titanium alloy matrix have been measured experimentally. All other thermoelastic constants of the matrix, together with all thermoelastic constants of the fibres are assumed to be independent of the temperature. The predicted thermal expansion data for UD composites, together with estimates of other thermoelastic constants, are then used to predict the thermal expansion coefficients of crossply laminates using classical laminate theory.

The predictions for the thermal expansion coefficients of unidirectional composites and cross-ply laminates are compared to experimental data and shown to be in very good agreement thus establishing the required validation. The models are used to investigate the effects of a carbon fibre coating on the predictions of thermal expansion coefficients. An approximate model of fibre/matrix debonding is also discussed.

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1 Introduction

Fibre reinforced titanium alloy composites are being evaluated for application in aeroengine components where the materials will be subjected to variations in temperature as well as variations in loading. The dimensional stability of components made from such materials is one feature of performance that must be considered, especially when there is selective reinforcement. Dimensional stability for uniform temperature distributions is influenced by the thermal expansion behaviour of the materials in the component. It is, therefore, vital to have information on the thermal expansion behaviour of fibre reinforced titanium composites. The objective of this report is to assess, by comparison with experimental results, selected methods of calculating the thermal expansion coefficients of unidirectional composites and cross-ply laminates from the properties of the silicon carbide fibre and of the titanium alloy matrix. In principle it is possible to take account of the temperature dependence of the properties of both fibres and matrix, although in this report consideration will be restricted to the temperature dependence of the Young's modulus and thermal expansion for the matrix.

The notation to be used will use the parameters E , ν and α to denote the Young's modulus, Poisson's ratio and the thermal expansion coefficient. Suffices f and m will denote that the thermoelastic constants refer respectively to the fibre and matrix, while the suffices A and T will denote the axial and transverse values for unidirectional and laminated composites. For unidirectional composites the axial direction clearly corresponds to the direction of the fibres, while for crossply laminates the axial direction is chosen to be that direction in which most fibres are oriented which usually corresponds to the principal loading direction. The superscript $*$ will denote that the composite property is defined by the rule of mixtures. An over bar will denote thermoelastic constants for a symmetric cross-ply laminate.

2 Model for unidirectional composites

Hashin (1983) has given a good review of theoretical models that can be used to predict the thermoelastic properties and thermal conductivities of composite materials. For unidirectional fibre reinforced composites, the concentric cylinder model has been used extensively to predict the thermoelastic constants in terms of the properties of the fibres and the matrix. Assuming that the fibres and matrix may both be assumed to be isotropic, the expressions for the axial and transverse thermal expansion coefficients are predicted by the concentric cylinders model to be (see McCartney (1989))

$$E_A \alpha_A = E_A^* \alpha_A^* + 2\lambda (\nu_m - \nu_f) V_f V_m [(1 + \nu_m) \alpha_m - (1 + \nu_f) \alpha_f], \quad (1)$$

$$\begin{aligned} \alpha_T = & -\nu_A \alpha_A + (1 + \nu_f) \alpha_f V_f + (1 + \nu_m) \alpha_m V_m \\ & + \frac{\lambda}{2} \left(\frac{1}{k_f} - \frac{1}{k_m} \right) V_f V_m [(1 + \nu_m) \alpha_m - (1 + \nu_f) \alpha_f], \end{aligned} \quad (2)$$

where

$$E_A = E_A^* + 2\lambda (\nu_m - \nu_f)^2 V_f V_m, \quad (3)$$

$$v_A = v_A^* - \frac{\lambda}{2} (v_f - v_m) \left(\frac{1}{k_f} - \frac{1}{k_m} \right) V_f V_m, \quad (4)$$

and where in (1-4)

$$E_A^* = V_f E_f + V_m E_m, \quad (5)$$

$$v_A^* = V_f v_f + V_m v_m, \quad (6)$$

$$E_A^* \alpha_A^* = V_f E_f \alpha_f + V_m E_m \alpha_m, \quad (7)$$

$$\frac{1}{\lambda} = \frac{1}{2} \left[\frac{1}{\mu_m} + \frac{V_f}{k_m} + \frac{V_m}{k_f} \right]. \quad (8)$$

The parameters k_f and k_m are the bulk moduli of fibre and matrix respectively defined by

$$k_f = \frac{E_f}{2(1 + v_f)(1 - 2v_f)}, \quad k_m = \frac{E_m}{2(1 + v_m)(1 - 2v_m)}. \quad (9)$$

The relations (1) and (2) are entirely equivalent to those given by Rosen and Hashin (1970), and Hashin (1983). The formulae (1) and (2) are, however, preferred as they have the same desirable structure as the relations (3) and (4) presented by Hashin (1983). The equations (1-9) can be generalised to fibres and matrix that behave as transverse isotropic materials, and to multiple systems of concentric cylinders (see McCartney (1992a)) so that the effects of fibre structure and coatings can be investigated.

3 Model for crossply laminates

Consider an undamaged multi-layered cross-ply laminate subject to biaxial in-plane loading. It is assumed that the external faces of the laminate are stress free, and that the laminate is thin enough for the through-thickness stress component to be everywhere zero for an undamaged laminate. Classical laminate theory applies and it can be shown, for the case when both 0° and 90° plies have the same properties, that the thermoelastic constants of a symmetric laminate may be obtained from the relations

$$\bar{E}_A = A - \frac{B^2}{C}, \quad \bar{E}_T = C - \frac{B^2}{A}, \quad \bar{v}_A = \frac{B}{C}, \quad (10)$$

$$\bar{E}_A \bar{\alpha}_A = P - \bar{\nu}_A Q, \quad \bar{E}_T \bar{\alpha}_T = Q - (\bar{\nu}_A \bar{E}_T / \bar{E}_A) P, \quad (11)$$

where $\bar{E}_A$ and $\bar{E}_T$ are, respectively, the axial and transverse Young's moduli of the laminate as a whole, $\bar{\alpha}_A$ and $\bar{\alpha}_T$ are the axial and transverse thermal expansion coefficients for the laminate, and $\bar{\nu}_A$ is the axial Poisson's ratio for the laminate. The parameters A, B, C, P and Q are defined by

$$A = \frac{1}{h} (b \hat{E}_A + a \hat{E}_T), \quad B = \nu_A \hat{E}_T, \quad C = \frac{1}{h} (b \hat{E}_T + a \hat{E}_A), \quad (12)$$

$$P = \frac{1}{h} (b \hat{E}_A \hat{\alpha}_A + a \hat{E}_T \hat{\alpha}_T), \quad Q = \frac{1}{h} (b \hat{E}_T \hat{\alpha}_T + a \hat{E}_A \hat{\alpha}_A), \quad (13)$$

where 2a is the total thickness of the 90° plies, 2b is the total thickness of the 0° plies, with $h = a + b$ so that 2h is the total thickness of the laminate, and where

$$\frac{1}{\hat{E}_A} = \frac{1}{E_A} \left(1 - (\nu_A)^2 \frac{E_T}{E_A} \right), \quad \hat{\alpha}_A = \alpha_A + \nu_A \frac{E_T}{E_A} \alpha_T, \quad (14)$$

$$\frac{1}{\hat{E}_T} = \frac{1}{E_T} \left(1 - (\nu_A)^2 \frac{E_T}{E_A} \right), \quad \hat{\alpha}_T = \alpha_T + \nu_A \alpha_A. \quad (15)$$

The relations (10-15) are a special case of known expressions (see McCartney (1992b)) for the thermoelastic constants of a crossply laminate where the 0° and 90° plies can have different properties.

It is seen from (14) and (15) that the value of transverse Young's modulus of the individual plies is needed, a parameter that was not specified in section 2. To calculate E_T it is first necessary to calculate the transverse shear modulus using the following formula (see for example McCartney (1989))

$$\frac{1}{k_T} \equiv \frac{2(1 - \nu_T)}{E_T} - \frac{4\nu_A^2}{E_A} = \frac{1}{k_T^*} - \frac{\lambda}{2} \left(\frac{1}{k_f} - \frac{1}{k_m} \right)^2 \nu_f \nu_m, \quad (16)$$

where

$$\frac{1}{k_T^*} = \frac{V_f}{k_f} + \frac{V_m}{k_m} . \quad (17)$$

As the values of E_A and v_A are known from (3) and (4) the value of $(1 - v_T)/E_T$ can now be calculated using (16). As the unidirectional composite is being modelled as a transverse isotropic solid the following relationship holds

$$E_T = 2\mu_T(1 + v_T), \quad \text{but} \quad E_A \neq 2\mu_A(1 + v_A), \quad (18)$$

where μ denotes the shear modulus. Clearly both E_T and v_T can be calculated once the value of μ_T has been determined. Christensen and Lo (1979) have developed a concentric cylinders model that can be used to estimate a value of the transverse shear modulus which lies between the bounds given by Hashin (1983), and likely therefore to predict more accurate values. The Christensen and Lo model has, therefore, been used to calculate the transverse shear modulus μ_T enabling values of E_T to be determined as described above. The details of the calculations are too complicated to be included here. The following expression given by Hashin (1983), leads to either an upper or lower bound value for μ_T , depending upon the values of μ_f , μ_m , k_f and k_m

$$\mu_T = \mu_m + \frac{V_f}{\frac{1}{\mu_f - \mu_m} + \frac{k_m + 2\mu_m}{2\mu_m(k_m + \mu_m)} V_m} . \quad (19)$$

4 Fibre properties

The properties for the silicon carbide fibres, assumed to be isotropic and independent of temperature, are given with sources in Table 1.

Table 1 - Fibre data from the literature

	E_f , GPa	v_f^*	α_f^{**} , $^{\circ}\text{C}^{-1}$
Sigma fibre	409	0.25	5×10^{-6}

* assumed by Sun et al. (1993)

** compromise between $4.86 \times 10^{-6} \text{ } ^{\circ}\text{C}^{-1}$ assumed by Nimmer et al. (1991) and $5.04 \times 10^{-6} \text{ } ^{\circ}\text{C}^{-1}$ assumed by Sun et al. (1993)

5 Matrix properties

The Young's modulus for the composite matrix has been measured at NPL for a range of temperatures between 25°C and 600°C using the impulse excitation technique (Wolfenden et al 1989). Moduli have been measured from a testpiece cut from a plate supplied by Rolls-Royce (see Acknowledgement) which had been prepared by consolidating 18 foils to a thickness of approximately 2.5 mm using the same manufacturing process as for the composites that have been tested. The results are shown in Table 2 and are plotted in Fig.1.

Table 2 - Measured matrix modulus data

Temperature, °C	Young's modulus GPa, (Run 1)	Young's modulus GPa, (Run 2)
25	116.4	116.9
80	114.8	
100	113.8	
200	108.7	107.4
250	106.4	104.6
300	103.3	102.0
350	101.1	99.3
400	98.0	97.1
400		97.5
450	94.8	94.5
500		90.9
500		91.6
550	90.2	88.0
600	88.1	85.2

It is seen from Fig.1 that the temperature dependence can be well represented by the following linear relationship

$$E_m(T) = \frac{86.65 (T - 25) + 116.65 (600 - T)}{600 - 25} .$$

This relation assumes that the temperatures are given in degrees Centigrade and that the moduli are in GPa. The linear relation is such that $E_m(25) = 116.65$ GPa and $E_m(600) = 86.65$ GPa. Nimmer et al (1991) have assumed (see Table 3) a room temperature modulus (at 21°C) of 113.7 GPa. This is lower than the value of 116.9 GPa predicted by the above linear relation and significantly higher than the value of 104.8 GPa that was used by Sun et al (1993) for Ti-6Al-4V consolidated from foils. Modulus data predicted by the above linear relation, that are to be used in the predictions of composite properties, are compared in Fig.2 with the data used by Nimmer et al (1991) (see Table 3) and a different linear relation used by Sun et al (1993).

Table 3 - Elastic moduli data for Ti-6Al-4V (Nimmer et al. (1991))

Temperature, °C	ν_m^*	E_m , GPa
21	0.34	113.7
149		107.5
315		97.9
482		81.3
649		49.6
900		20.7

* Assumed by Sun et al. (1993) for Ti-6-4 consolidated foils

Following Sun et al (1993) Poisson's ratio for the matrix is taken to be temperature independent, having the value 0.34. Assuming that the matrix elastic properties are isotropic, it is possible to calculate any other of the elastic constants for the matrix which will be temperature dependent because of the temperature dependence of Young's modulus.

The linear thermal expansion coefficient for the matrix has been measured for a range of temperatures between 21°C and 700°C using a modified calibrated Linseis silica dilatometer. Small strip test-pieces about 30 mm long and 8 mm wide were cut from consolidated titanium foils and from composites in orthogonal directions. Several thermal cycles at 2 °C/min heating and cooling rate were made on each test-piece to check uniformity and consistency of behaviour. Data were recorded on a data logger and the results in the form of expansion in parts per million as a function of temperature were calculated. From these data, expansivities ($d\ell/dT$) were estimated numerically.

The thermal expansion measurements for the Ti-6Al-4V matrix shown in Table 4 are the mean values for heating and cooling from test-pieces cut from the plate supplied by Rolls-Royce which had been prepared by consolidating 18 foils to a thickness of approximately 2.5 mm using the same manufacturing process as for the composites.

The dotted lines in Fig.3 show the thermal expansion measurements obtained in directions parallel and perpendicular to the rolling direction. The average data are shown in Fig.3 as a continuous line which tends to smooth the data. The experimental temperature dependent data for matrix modulus and thermal expansion coefficient are used in predictions for any temperature by linearly interpolating between the experimental data points. As a consequence theoretical predictions will not be smooth.

Table 4 - Thermal expansion data for Ti-6Al-4V

Temperature °C	Expansion coefficient $\alpha_m, 10^{-6} \text{ } ^\circ\text{C}^{-1}$
21	8.4 (extrapolated)
75	9.0
125	9.6
175	9.8
225	10.2
275	10.4
325	10.6
375	10.7
425	10.9
475	11.0
525	11.3
575	10.5
625	11.8

6 Experimental thermal expansion data for composites

The results of the thermal expansion measurements on both unidirectional fibre reinforced titanium and cross-ply laminates are shown in Figs. 4 and 5. In Fig.4 the axial and transverse thermal expansion coefficients for unidirectional material are given as a function of temperature. The material tested has a fibre volume fraction in the range 0.32 - 0.34. The value $V_f = 0.32$ has been used in all predictions. Fig.5 shows the thermal expansion coefficient for a symmetric cross-ply laminate having the same number of 0° and 90° plies so that the axial and transverse expansion coefficients would be expected to have the same value.

7 Comparison of models with experimental data

The relationships described in section 2 have been used to predict the temperature dependence of the axial and transverse thermal expansion coefficients of a unidirectional composite having a fibre volume fraction of 0.32. The results of these predictions, which are given in Fig.6, compare well with those measured experimentally, indicating that the values used for fibre and matrix properties are reasonable. It is noted that the measurement of thermal expansion coefficients at room temperature will be subject to a larger error than at higher temperatures because of differences that arise in the temperature distribution, and because no measurements have been made at lower temperatures. Also, at temperatures approaching 600°C , phase changes in the matrix lead to changes in expansion behaviour. In the temperature range $T > 300^\circ\text{C}$ the calculated values for the transverse modulus of unidirectional material, are above the measured values. This difference could be caused by interface debonding: a factor that will be considered later.

The calculated unidirectional properties of the composite have been used in conjunction with the relations (10) and (11) to calculate the thermal expansion coefficient of a crossply laminate where there are equal numbers of 0° and 90° plies and where in consequence $a = b$ so that the axial and transverse expansion coefficients of the laminate have the same value. The results of calculations are shown in Fig.6 together with the experimental results, indicating very good agreement for a wide temperature range.

8 Effect of a fibre coating

Tamin et al (1994) have included carbon coatings on fibres in their modelling of the time dependent behaviour of fibre reinforced metal matrix composites. They assumed that the coating was isotropic having a Young's modulus of 9.12 GPa and a thermal expansion coefficient of $2.427 \times 10^{-6} / ^\circ\text{C}$. Robertson (1996) has suggested that for DRA Sigma composites the axial and hoop moduli for carbon coatings are in the range 100-120 GPa and that the radial modulus of the coating is less than 100 GPa. For predictions it has, therefore, been assumed that the carbon coating is transverse isotropic such the axial Young's modulus is 110 GPa and the transverse Young's modulus is 100 GPa. The thermal expansion coefficient for the carbon coating has been assumed to be isotropic having the value above used by Tamin et al (1994). It is assumed that Poisson's ratio for the carbon coating has the value 0.3. Calculations have been performed including a carbon coating of thickness $t = 4.5 \mu\text{m}$ (Robertson (1996)) for a fibre of diameter $100 \mu\text{m}$ such that $t/R = 0.09$ where R is the fibre radius. The predictive methods used are given in McCartney (1992a). The results of including the carbon coatings are shown in Fig.7 where theoretical predictions are compared with the experimental results. By comparing Figs.6 and 7 it is clear that the presence of the coating has improved the predictions of the axial thermal expansion coefficient for the unidirectional composite, but led to a greater discrepancy between predicted and experimental values for the transverse thermal expansion coefficient. The predictions for the cross-ply laminate are also improved by accounting for the presence of the coating. It is worth noting that the discrepancy between predicted and experimental values for the transverse expansion coefficient could arise from the assumption that the thermal expansion coefficient for the coating is isotropic.

9 Effect of interface debonding

In an attempt to determine the effect of fibre/matrix debonding, the Young's modulus and thermal expansion coefficient of the coating were set to very small values (Young's modulus = 0.001 GPa, expansion coefficient = $1 \times 10^{-12} / ^\circ\text{C}$) compared to those of the fibre and matrix. Poisson's ratio for the interlayer was again assumed to be 0.3. The thickness t of the interlayer between fibre and matrix was selected so that $t/R = 0.001$ where R is the fibre radius. One characteristic of model is that there is negligible transfer of stress between fibre and matrix across the interlayer. Because the model assumes perfect bonding between fibre, interlayer and matrix the resulting predictions will exhibit some unrealistic features such as the apparent inter-penetration of fibre and matrix material in the neighbourhood of the interlayer. For a debonded fibre/matrix interface there would be some stress transfer between fibre and matrix whenever a ply is subject to transverse loading. It is concluded that, while the model predictions are known to be inaccurate, they will indicate exaggerated trends in behaviour. The effects of fibre/matrix debonding are shown in Fig.8 where, by comparison

with Fig.6, it is very clear that the transverse thermal expansion coefficient of unidirectionally reinforced composite increases significantly when compared to predictions for perfectly bonded interfaces. This might explain the experimental observation that the measured values of the transverse thermal expansion coefficient of unidirectional material rises above the predictions when the temperature is greater than 300°C. It may also explain why the expansion coefficient does not fall at 500°C as might be expected from the matrix data shown in Fig.3. It is interesting to note that the thermal expansion of a crossply laminate is hardly affected by fibre/matrix debonding. Figs.9-11 compare theoretical predictions of the thermal expansion coefficients for two-phase composites, composites with carbon coated fibres and composites with fully debonded fibre/matrix interfaces.

10 Conclusions

It is concluded that the models suggested for the calculation of the thermal expansion coefficients of silicon carbide fibre reinforced titanium alloy are giving results that are in very good agreement with experimentally measured values. It is also concluded that the values used for the properties of the fibres and matrix appear to be satisfactory for application in the models. Predictions for a model where carbon fibre coatings are included indicate that they are expected to have minimal effect on the axial thermal expansion coefficient of unidirectional composites and of crossply laminates provided that the interface remains perfectly bonded. A more significant effect is expected on the transverse expansion coefficient for unidirectional composites. An approximate analysis of fibre/matrix debonding indicates that only the transverse thermal expansion coefficient of unidirectional material is likely to be affected significantly as to be expected.

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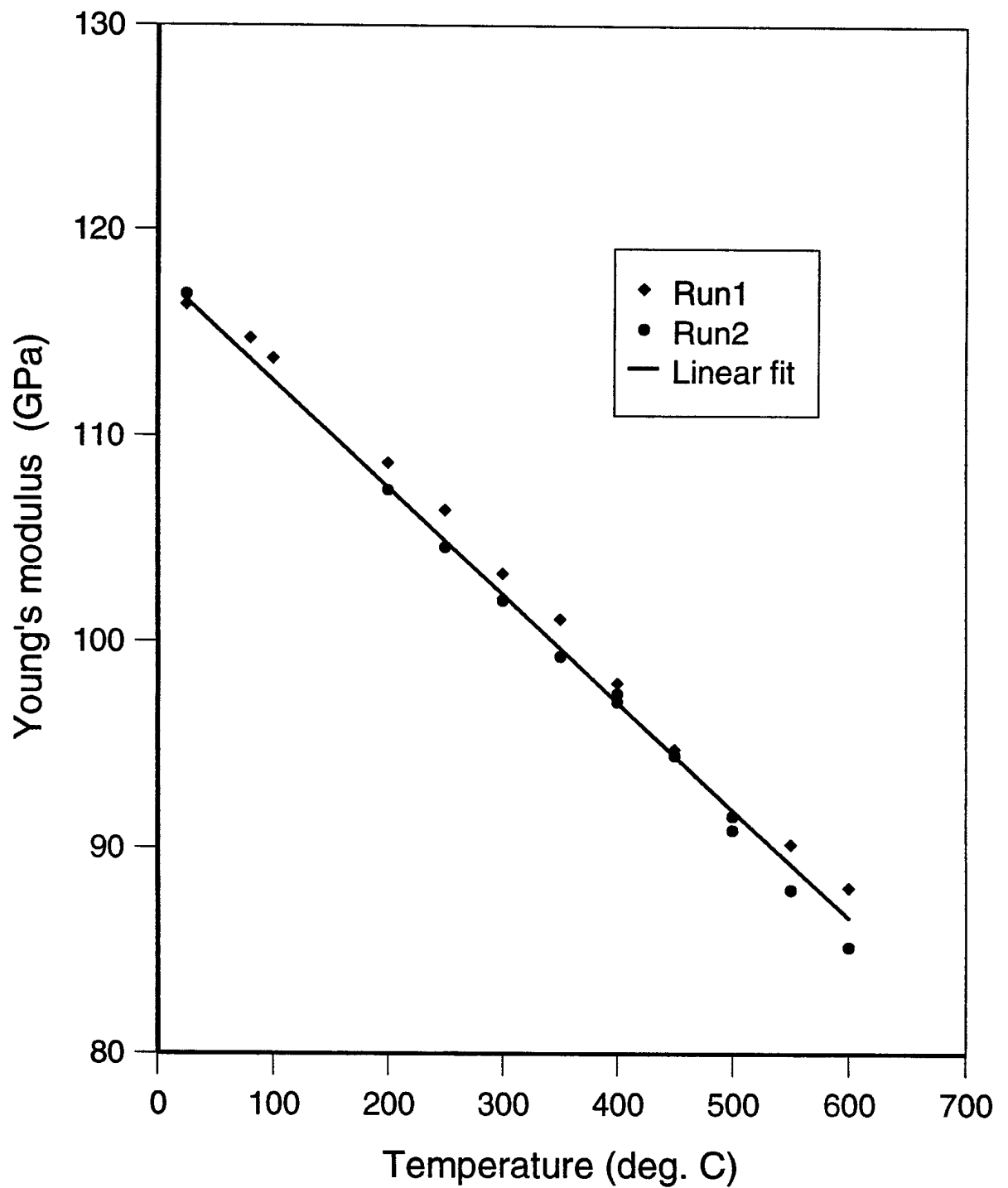


FIG.1 : Dependence of Young's modulus for laminated Ti - 6Al - 4V alloy on temperature, showing experimental impulse excitation data as discrete points and the fitted linear approximation.

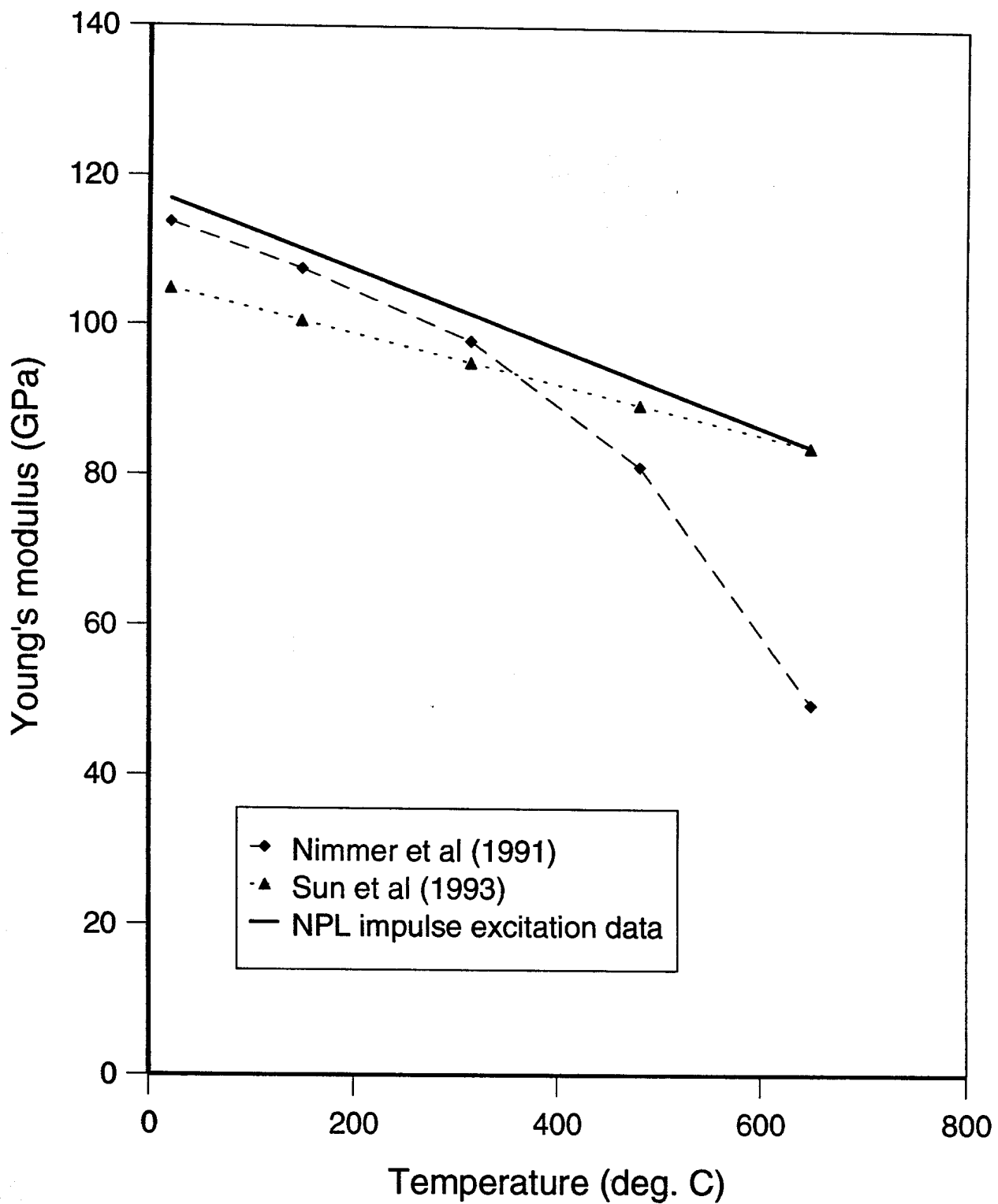


FIG.2 : Temperature dependence of Young's modulus for Ti-6Al-4V titanium alloy

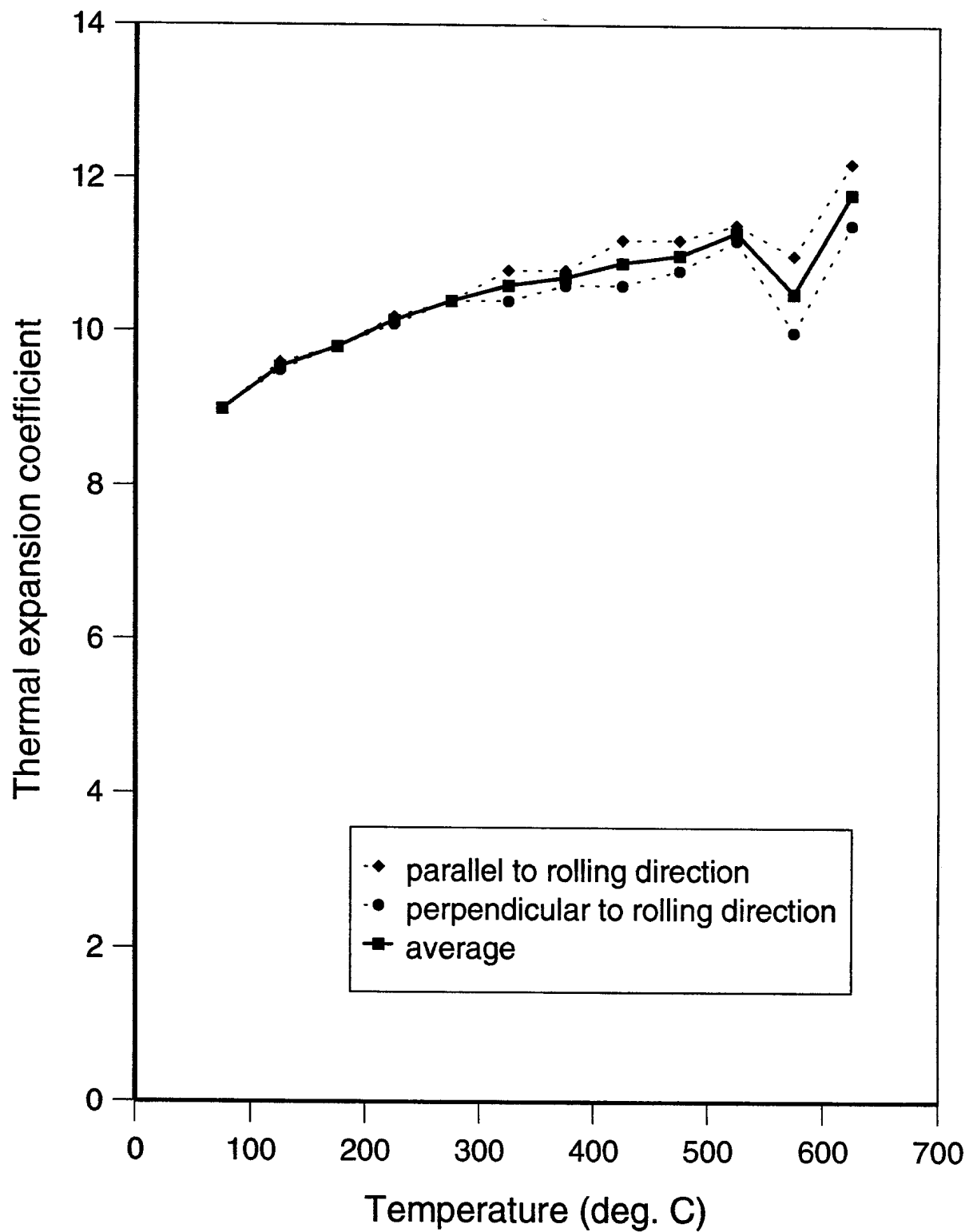


FIG.3 : Experimentally measured values of the thermal expansion coefficient (millionths) per degree C for Ti-6Al-4V titanium alloy.

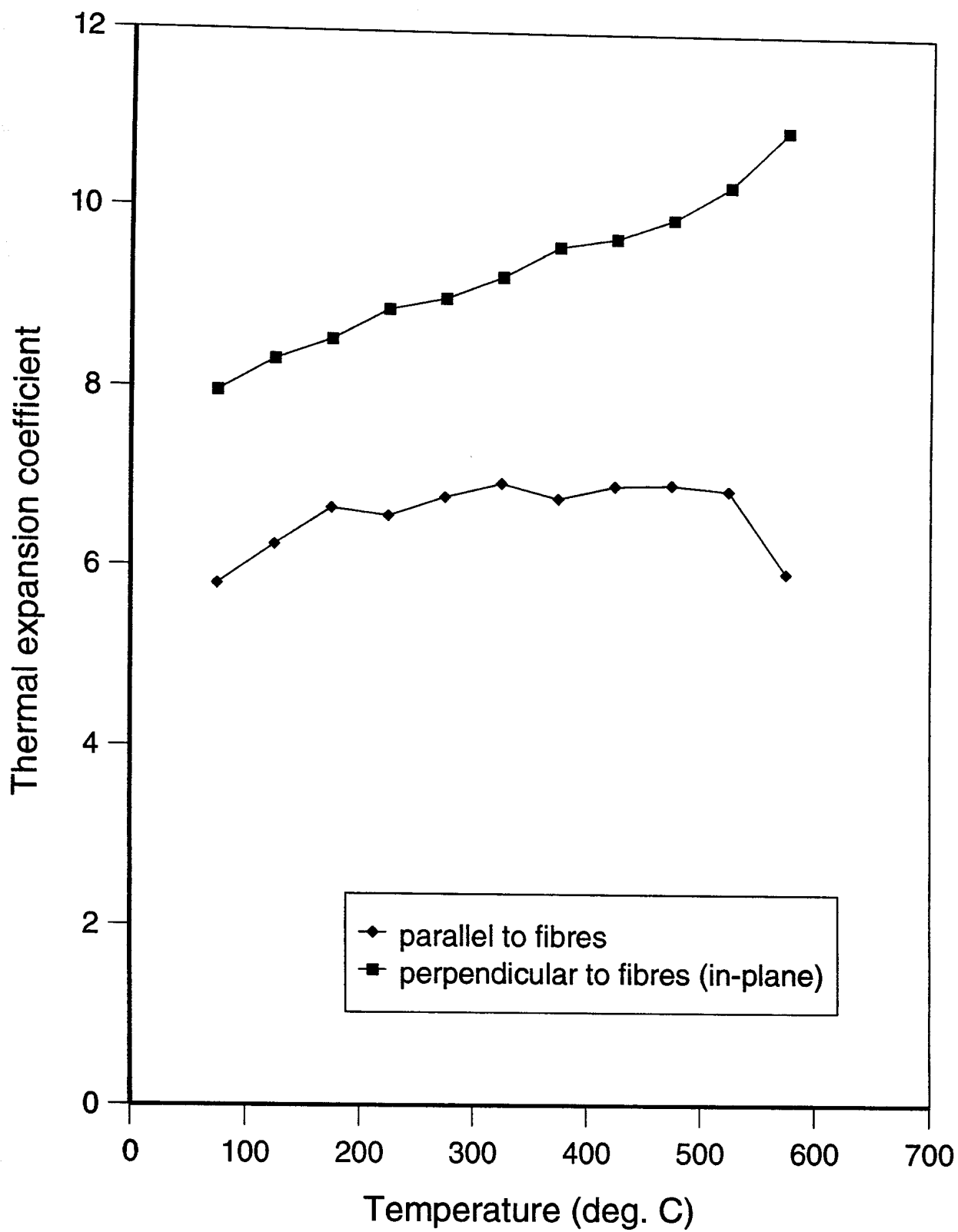


Fig.4 : Experimentally measured values of the thermal expansion coefficients (millionths) per degree C of unidirectional Sigma silicon carbide fibre reinforced Ti-6Al-4V composite.

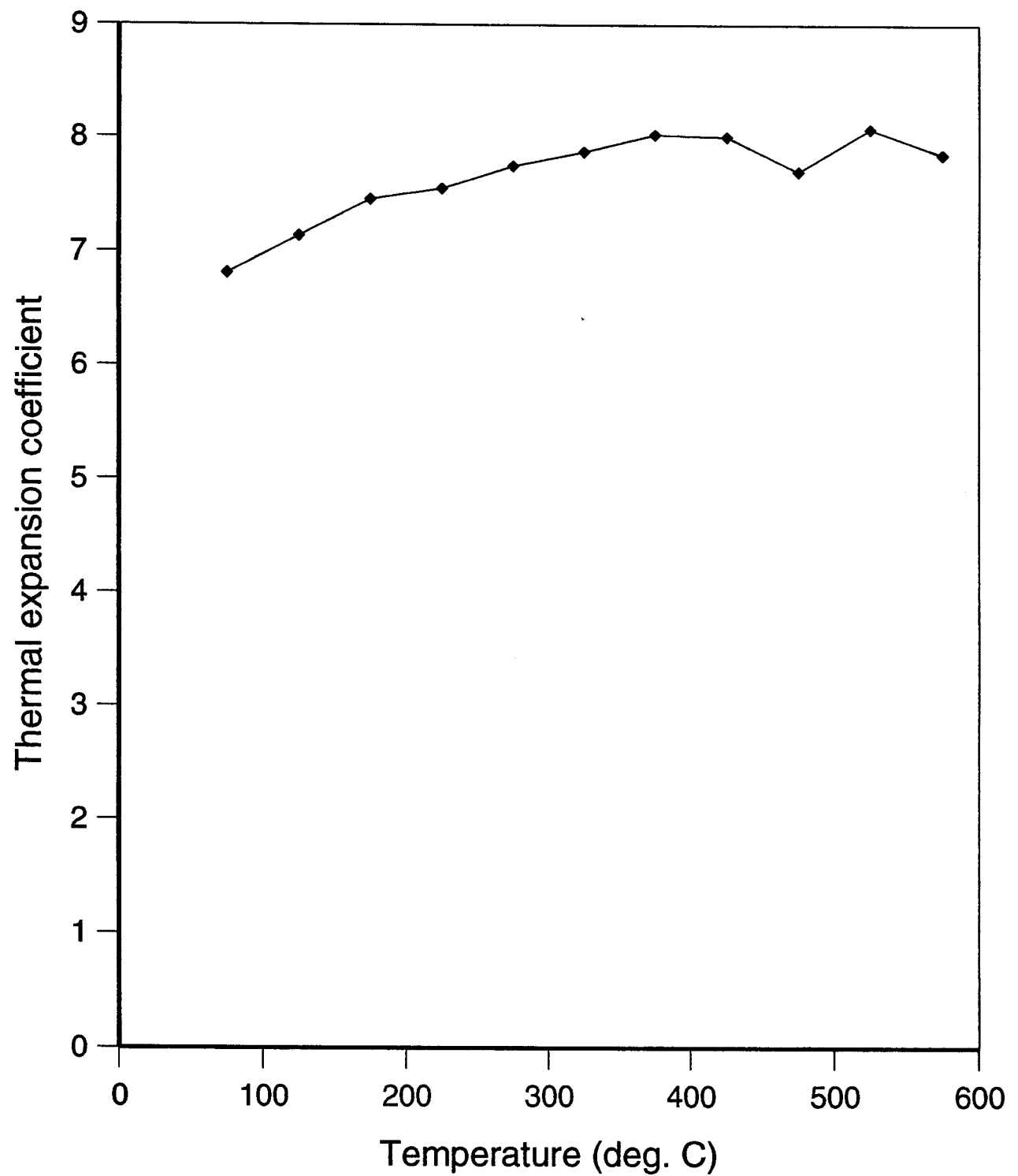


FIG.5 : Experimentally measured values for the thermal expansion (millionths) per degree C of a crossply laminate made from Sigma silicon carbide fibres in a Ti-6Al-4V matrix.

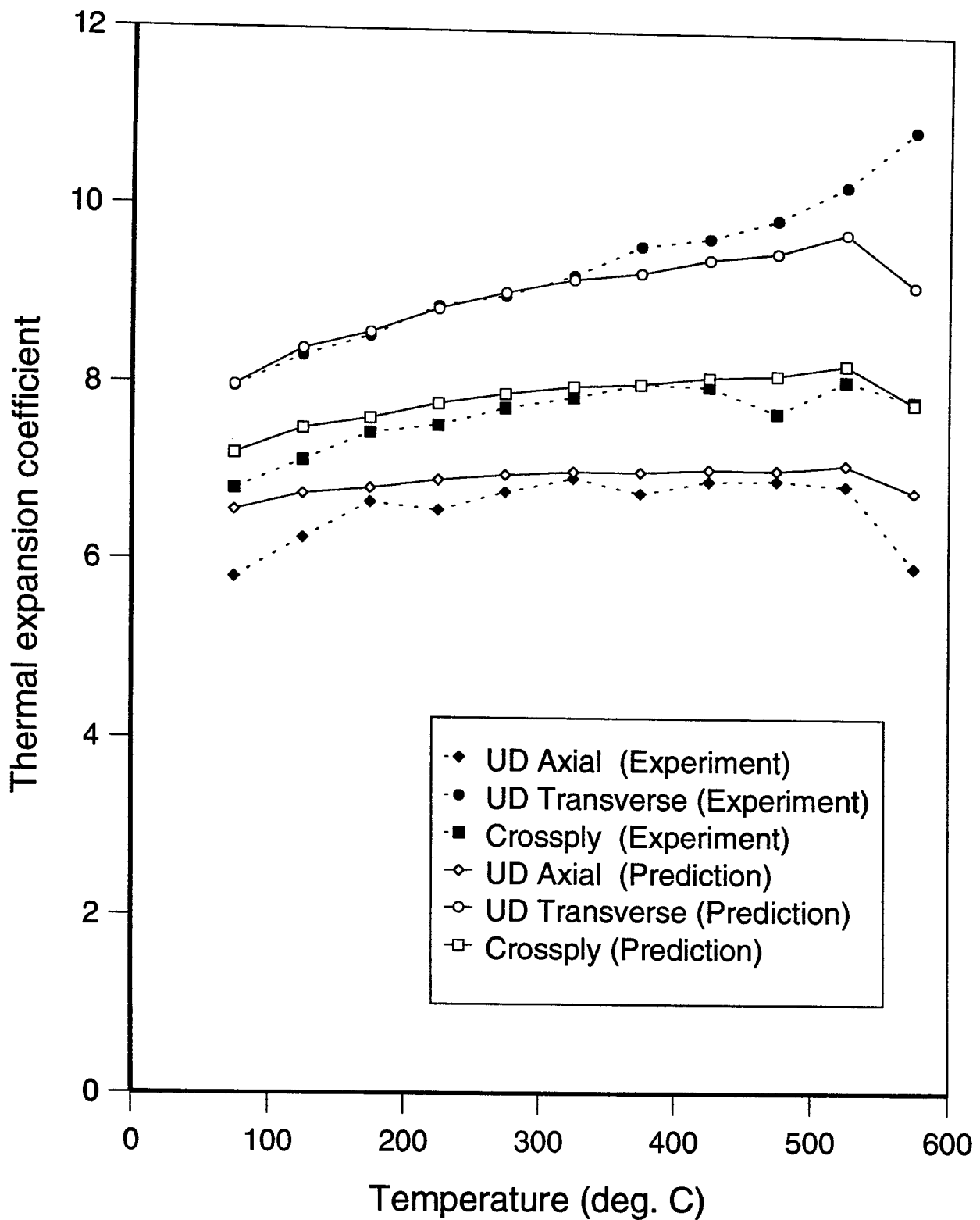


FIG.6 : Comparison with experimental data of model predictions for the thermal expansion coefficients (millionths) per degree C of Sigma silicon carbide fibre reinforced Ti-6Al-4V titanium composites.

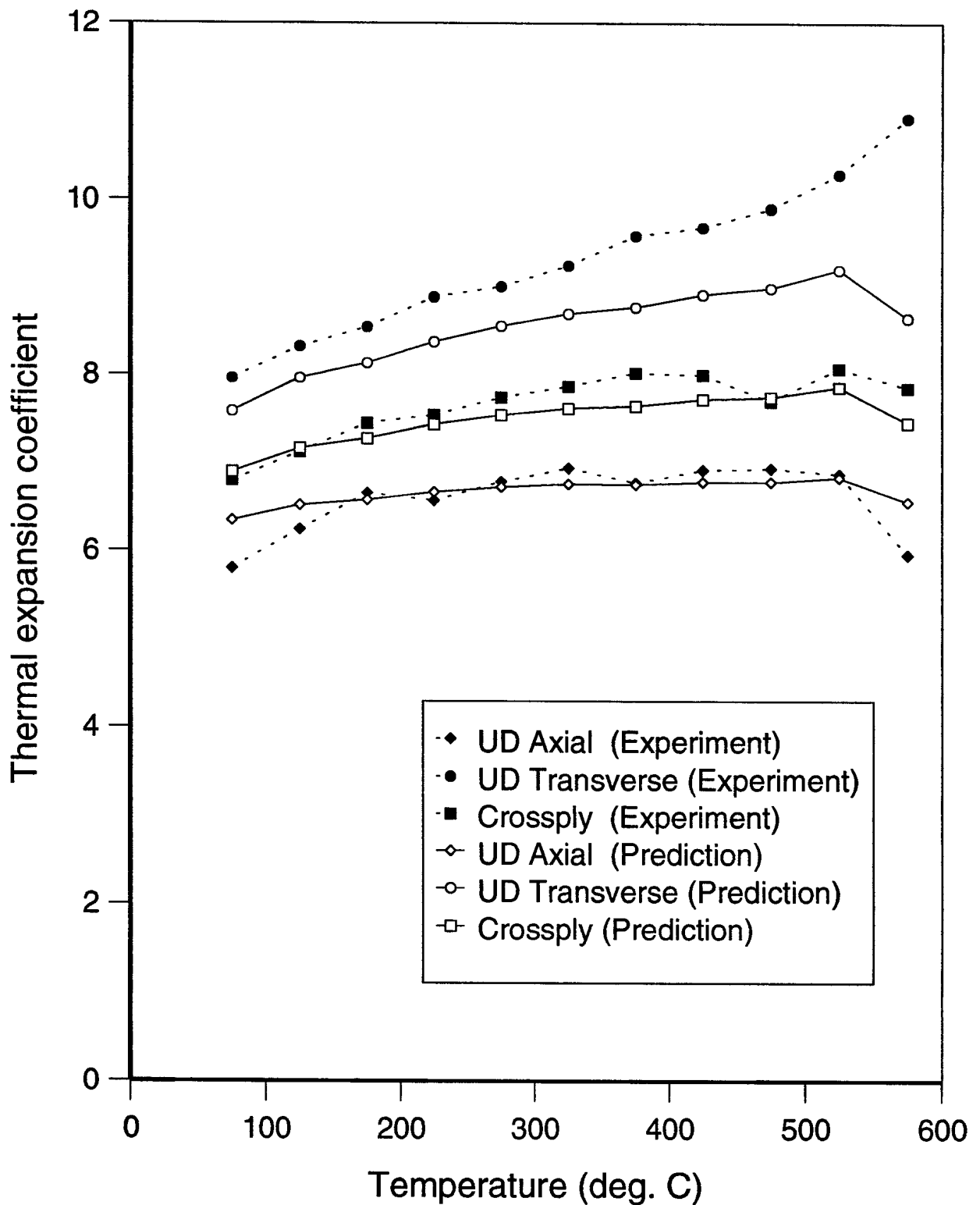


FIG.7 : Comparison with experimental data of model predictions for the thermal expansion coefficients (millionths) per degree C of Sigma silicon carbide fibre reinforced Ti-6Al-4V titanium composites for case when there is a coating at fibre/matrix interface.

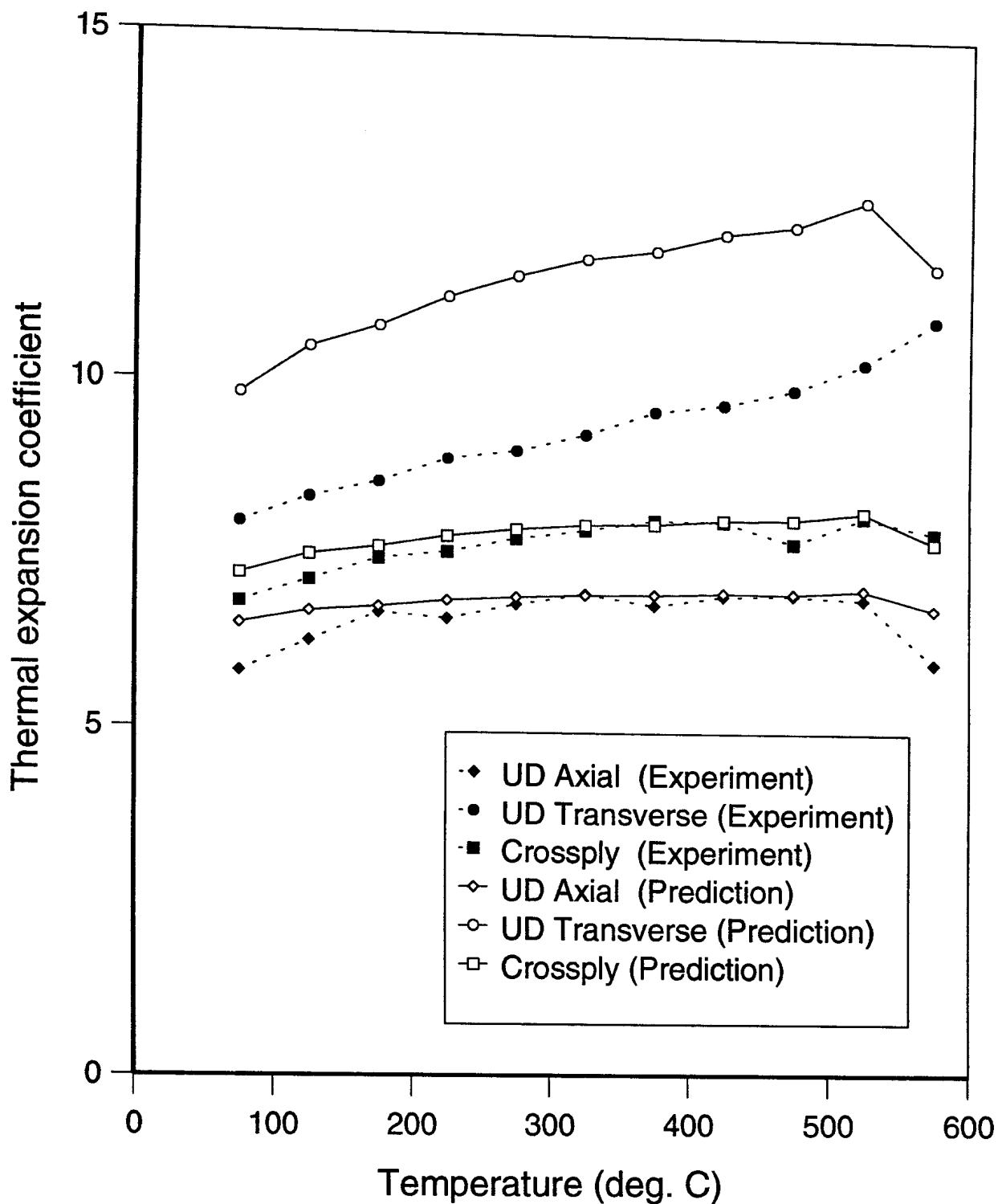


FIG.8 : Comparison with experimental data of model predictions for the thermal expansion coefficients (millionths) per degree C of Sigma silicon carbide fibre reinforced Ti-6Al-4V titanium composites for the case when there is a debonded fibre/matrix interface.

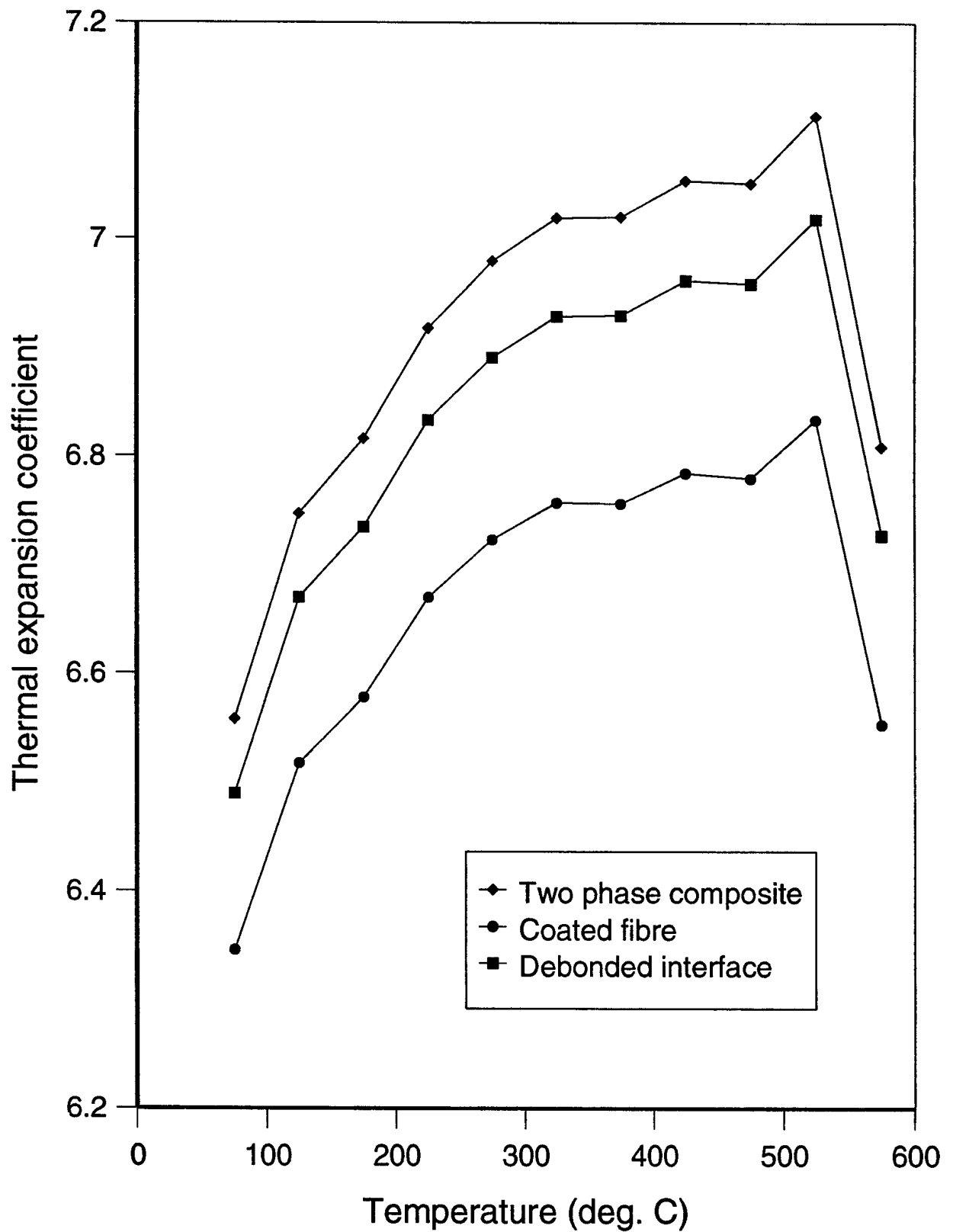


Fig.9 : Predictions of the axial thermal expansion coefficient (millionths) per degree C for a unidirectional fibre reinforced titanium composite for three different interfacial conditions.

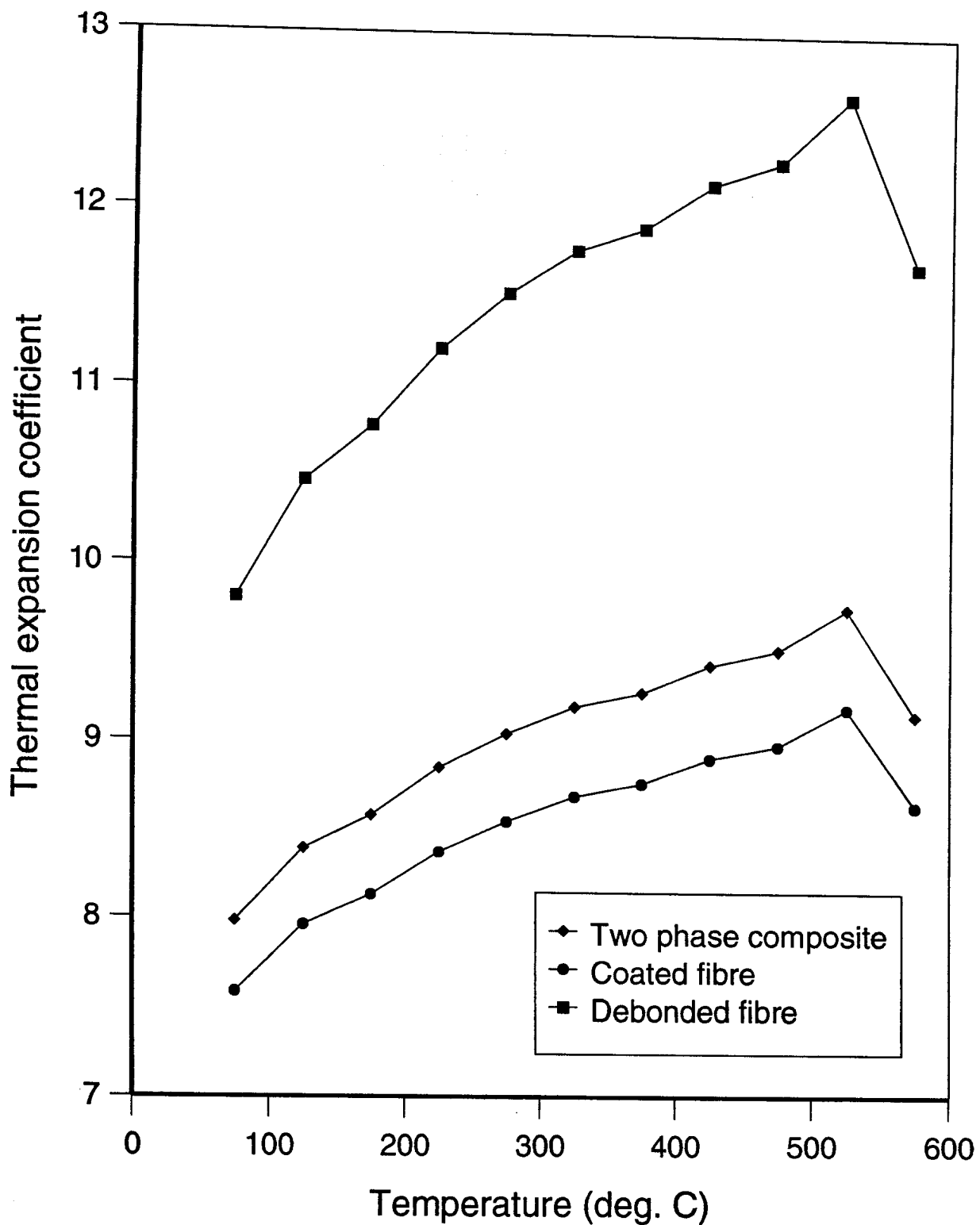


Fig.10 : Predictions of the transverse thermal expansion coefficient (millionths) per degree C for a unidirectional fibre reinforced titanium composite for three different interfacial conditions.

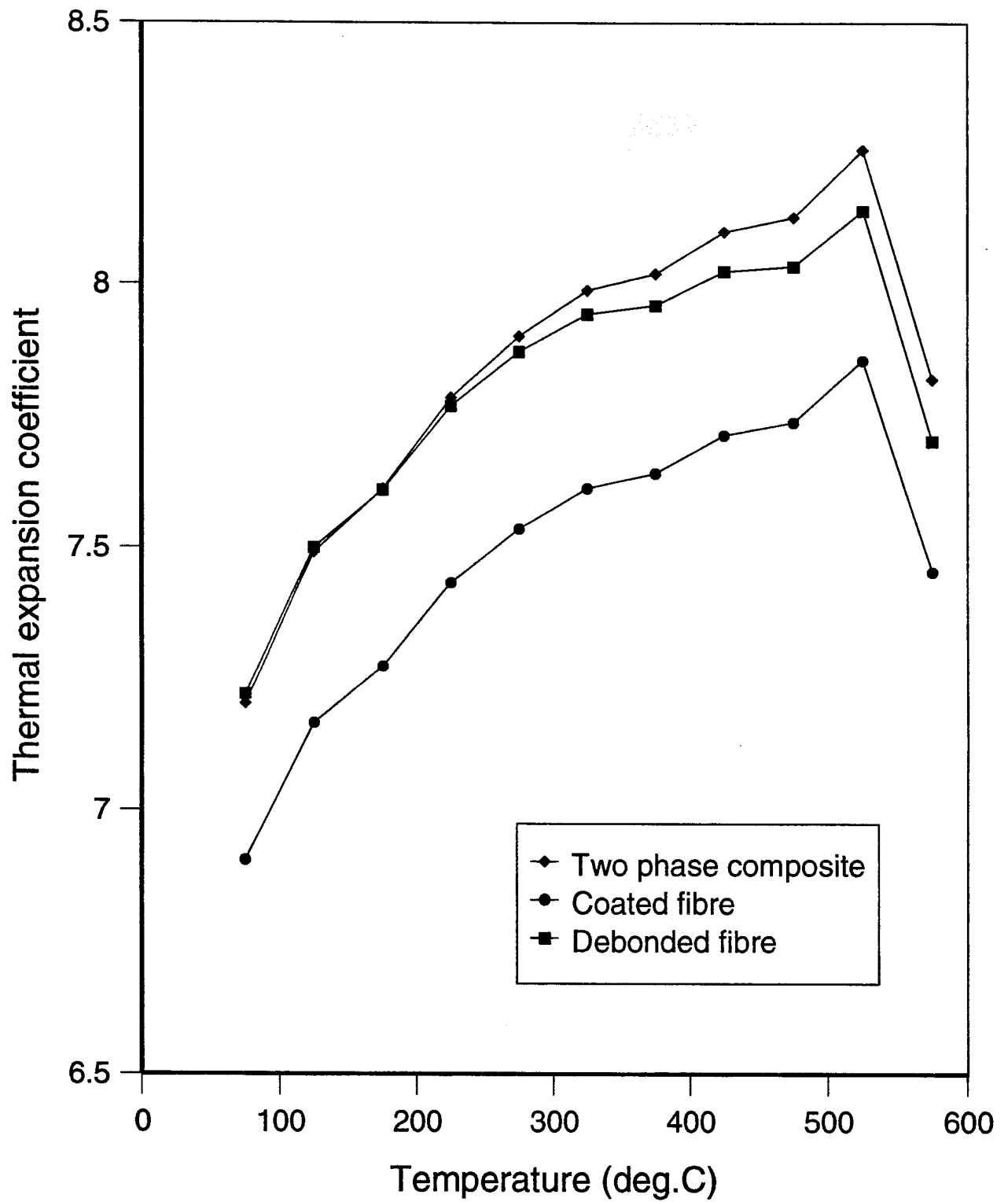


Fig.11 : Predictions of thermal expansion coefficient (millionths) per degree C for a cross-ply titanium composite for three different interfacial conditions.