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The fracture behaviour of two Co–Mo–Cr–Si wear resistant alloys ("Tribaloys")

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The room temperature modulus of rupture (MOR) and the plane strain fracture toughness (K_{IC}) of two Co–Mo–Cr–Si wear resistant alloys in various microstructure conditions have been determined. The modulus of rupture varied from 304 to 927 MN m^{-2} and as such was more structure sensitive than K_{IC}, which was in the range 15 to 26 MN m$^{-3/2}$ for all conditions of the alloys. The best combination of room-temperature properties was associated with an f c c solid solution whereas h c p solid solution with a Widmanstätten precipitate produced the poorest properties. The industrial significance of these results are discussed.

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1. Introduction

The "Tribaloys"† are a group of water-resistant alloys that gain their strength and hardness from a coarse dispersion of particles of an intermetallic Laves phase. The compositions of the two cobalt "Tribaloys", T400 and T800 are given in Table I. The main differences between T400 and T800 are the greater amount of intermetallic phase and higher chromium content in T800, which results in a harder alloy that retains excellent wear resistance in more corrosive environments. In common with other commercial wear-resistant alloys, T400 and T800 are hard, have high bearing strengths but exhibit little capacity for plastic flow. The basic mechanical property data [1] for these alloys are presented in Table II.

These cobalt alloys were believed [1–3] to be stable up to at least 1230° C and it was considered that the microstructure could not be modified by subsequent heat treatment after casting. However, recent work [4, 5] has shown this to be incorrect and the alloys can, in fact, be hardened or softened by heat treatment.

In the as-cast condition, the "Tribaloys" (Figs. 1 and 2) both have large primary dendrites of the hard intermetallic phase, which is the Laves phase Mo(Co, Si)$_2$ of the MgZn$_2$ type. The primary dendrites are in a eutectic matrix of smaller intermetallic particles embedded in a cobalt solid solution; there are also regions of cobalt solid solution which are free from the secondary Laves phase. The allotropic nature of cobalt can cause either the face centred cubic (f c c) or hexagonal close packed (h c p) crystal structures or both to be present in "Tribaloys" depending on the thermal treatment. In addition by heat treatment in the range 600 to 1100° C it is possible to obtain precipitation within the cobalt solid solution in either a spheroidal or a Widmanstätten form.

The effect of precipitation and the allotropic transformation on the hardness has been discussed in a previous publication [5]. This paper reports the results of an investigation of the role of the microstructure in determining the fracture behaviour. Casting methods and heat treatments were chosen to obtain a range of microstructural conditions, the main features of which are listed in Table III.

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TABLE I Nominal composition (wt %) and vol % of Laves phases [1]

Tribaloy	Co	Mo	Cr	Si	C	Vol % Laves
T-400	61–63	27–29	7.5– 8.5	2.2–2.6	less 0.08	40–50
T-800	51–53	27–29	16.5–17.5	3.0–3.5	less 0.08	40–60

2. Experimental procedure

Compression tests were carried out in order to obtain a value for the yield stress, σ_Y. This was necessary to ensure that the thickness (B) of the fracture toughness (K_{IC}) specimens was sufficient for plane strain conditions i.e. $B \geqslant 2.5\ (K_{IC}/\sigma_Y)^2$. Specimens of nominal size $37 \times 5 \times 2.5\ \text{mm}^3$ were then prepared from T800 and T400 in the various microstructural conditions. Some of these specimens were polished to a 1 μm diamond paste finish and loaded to failure in three-point bending using an Instron. The modulus of rupture (MOR) was calculated from the failure load (P) using the following equation:

$$\text{MOR} = \frac{3PL}{2BW^{3/2}} \tag{1}$$

where B and W are the specimen thickness and width respectively and L is the loading span.

Single-edge notch bend (SENB) specimens were produced from the remaining specimens. First a notch of depth $\sim \frac{1}{3}\ W$ was introduced into the specimen by means of a fine SiC slitting wheel. For satisfactory fracture toughness determination this notch has to be sharpened by some pre-cracking procedure which is normally that of fatigue. However, in hard metals the stress required to propagate a fatigue crack is so near the critical stress that the method was found to be unreliable because of the catastrophic failure that generally resulted. Other techniques in general use for brittle material, indentation [6] and wedge loading [7] were tried but the most successful and hence the one eventually used throughout the investigation, was a new method based on acoustic emission monitoring. Previous work on ceramics has demonstrated that acoustic emission was sensitive enough to detect sub-critical crack growth [8]. This suggested that sub-critical crack growth during slow loading of the SENB "Tribaloy" specimens could be quantified by acoustic emission and that rapid unloading before failure would produce sharpened cracks suitable for fracture toughness determination. For most of the microstructural conditions of the two alloys, it was found that a ringdown count rate over 90 000 sec^{-1} was required to extend the notch to a crack of the required length ~ 0.5 W. As shown in Fig. 3 the load required for this amount of crack extension was variable and not such a reliable guide as the acoustic emission for rapid unloading. An alternative acoustic emission technique was used for the alloys in the harder heat-treated conditions. These alloys were pre-cracked by incremental loading until a load level was attained at which emissions continued whilst the cross-head was stationary (Fig. 4).

TABLE II Typical mechanical properties of Tribaloy [1]

Property	T-400	T-800
Tensile strength (MN m^{-2})	690	–
Compressive strength (MN m^{-2})	1896	1780
Modulus of elasticity (GN m^{-2})	266	243
Charpy impact strength [un-notched (J)]	4.1	1.4

The pre-cracked specimens were tested in a similar manner to the modulus of rupture specimens and the critical stress intensity factor (K_{IC}) was evaluated from:

$$K_{IC} = \frac{3PLYa^{1/2}}{2BW^2}. \tag{2}$$

Y is a compliance factor to account for the variation in a/W, in this instance it is given by Brown and Strawley [9] for $L/W = 4$:

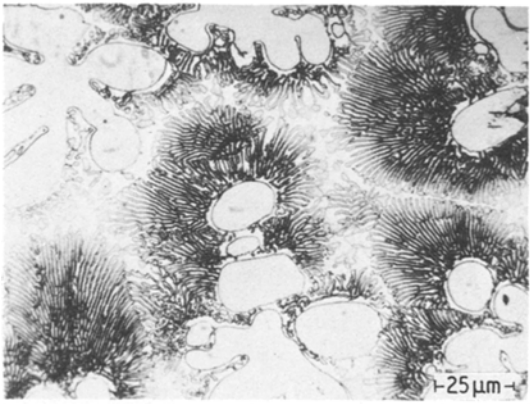

Figure 1 T-400 as-cast microstructure: primary Laves phase surrounded by cobalt solid solution and eutectic.

TABLE III Microstructural condition of the specimens under test

No.	Specimen condition	Volume fraction Laves	Mean size primary Laves	% lamellar eutectic	Cobalt solid solution			Precipitate present	Precipitate morphology
					%	fcc	hcp		
1	T-400 as commercially cast – fairly slow cool	33 ± 4	24 ± 20	47 ± 5	20 ± 5	√	√	No	
2	T-400 arc melted – rapid cooling	42 ± 8	6 ± 7	58 ± 11		√	trace	No	
3	T-400 as no. 1 followed by 2 h at 1250° C water quenched	41 ± 7	25 ± 3	29 ± 4	33 ± 3	√	–	No	–
4	T-400 as no. 3 followed by 13 days at 800° C water quenched	35 ± 4	17 ± 18	29 ± 4	33 ± 3	–	√	Yes	Widmanstätten
5	T-400 as no. 1 followed by 13 days at 800° C water quenched	32 ± 6	16 ± 8	50 ± 5	16 ± 3	trace	√	Yes	Spheroidal
6	T-800 as commercially cast – fairly slow cool	59 ± 8	25 ± 20	16 ± 5	25 ± 5	√	√	–	–
7	T-800 arc melted – rapid cooling	70 ± 12	9 ± 7	30 ± 5		√	trace		
8	T-800 as no. 6 followed by 2 h at 1250° C water quenched	53 ± 10	24 ± 10	10 ± 5	36 ± 3	√	–	–	–
9	T-800 as no. 8 followed by 13 days at 800° C water quenched	72 ± 6	28 ± 26	–	27 ± 4	–	√	Yes	Widmanstätten
10	T-800 as no. 6 followed by 13 days at 800° C water quenched	58 ± 7	27 ± 25	12 ± 5	32 ± 4	trace	√	Yes	Spheroidal

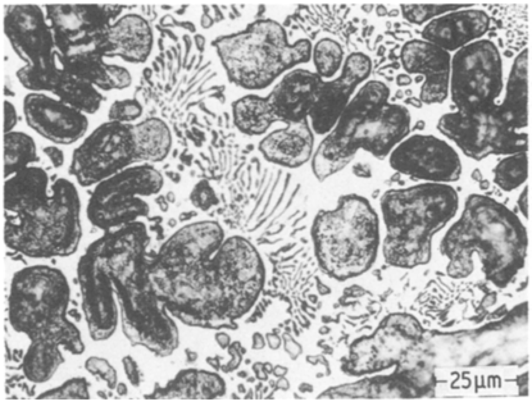

Figure 2 T-800 as-cast microstructure: primary Laves phase in a matrix of eutectic and Laves free solid solution.

$$Y = \left[1.93 - 3.07\left(\frac{a}{W}\right) + 14.53\left(\frac{a}{W}\right)^2 - 25.11\left(\frac{a}{W}\right)^3 + 25.80\left(\frac{a}{W}\right)^4\right].$$

When examining fracture in materials in the absence of a machined notch the general fracture equation:

$$\text{MOR} = \frac{K_{\text{IC}}}{Ya^{1/2}},$$

which is applicable to linear crack fronts, has to be modified to allow for the flaw shape by introducing a constant Z such that:

$$\text{MOR} = \frac{ZK_{\text{IC}}}{Ya^{1/2}} .$$

Bansal [10] demonstrated that for most elliptical flaws of practical significance $Z^2 = 2.82\,bA_{\text{c}}^{1/2}$ where A_{c} is the area of flaw and b is the semi major axis. This leads to the modification of the general fracture equation as:

$$\text{MOR} = \frac{1.68\,K_{\text{IC}}}{YA_{\text{c}}^{1/4}} \qquad (3)$$

The use of Equation 3 allows a critical defect size to be estimated using the experimentally determined values of K_{IC} and MOR. This provides a quantitative approach to the inherent flaw size within the material. The weakest material will always reflect the lowest combination of K_{IC} and MOR and this balance can be expressed as a large inherent flaw in the material.

3. Experimental results

Table IV shows the results of the fracture tests carried out on T400 and T800 in both the as-cast and heat-treated conditions. Also summarized in the table are the hardness and main microstructural

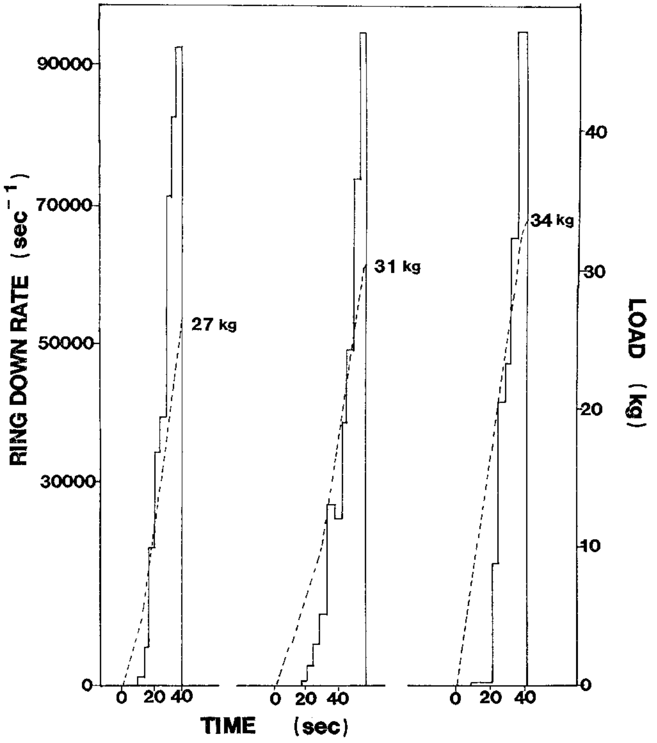

Figure 3 Pre-cracking fracture toughness specimens using the acoustic response. Identical ringdown response rates are observed when cracking occurs despite differences in terms of load.

TABLE IV Fracture toughness and strength data for T400 and T800 in various microstructural conditions.

No.	Specimen condition	Size primary Laves (μm)	Cobalt Solid solution	Precipitate morphology	K_{IC} (MN m$^{-3/2}$)	MOR (MN m^{-2})	Critical defect size (mm^2)	Hardness HV_{50}
1	T-400 as commercially cast	24	hcp/fcc	–	21.9 ± 3	745 ± 100	0.37	690 ± 22
2	T-400 arc melted	6	fcc/trace hcp	–	21.1 ± 3.4	917 ± 54	0.14	658 ± 25
3	T-400 as no. 1 + 2 h at 1250° C	25	fcc	–	26.0 ± 0.5	923 ± 54	0.31	620 ± 17
4	T-400 as no. 3 + 13 days at 800° C	17	hcp	Widmanstätten	15.8 ± 0.9	400 ± 100	1.21	736 ± 12
5	T-400 as no. 1 + 13 days at 800° C	16	hcp/trace fcc	Spheroidal	19.2 ± 0.8	506 ± 76	1.03	740 ± 12
6	T-800 as commercially cast	25	hcp/fcc	–	19.5 ± 1.5	187 ± 45	0.16	700 ± 20
7	T-800 arc melted	9	fcc/hcp	–	18.7 ± 1.5	752 ± 35	0.19	728 ± 15
8	T-800 as no. 6 + 2 h at 1250° C	24	fcc	–	24.2 ± 1.4	836 ± 114	0.35	650 ± 12
9	T-800 as no. 8 + 13 days at 800° C	28	hcp	Widmanstätten	18.7 ± 0.8	304 ± 97	7.13	816 ± 12
10	T-800 as no. 6 + 13 days at 800° C	27	hcp	Spheroidal	22.3 ± 1.7	583 ± 34	1.07	820 ± 17

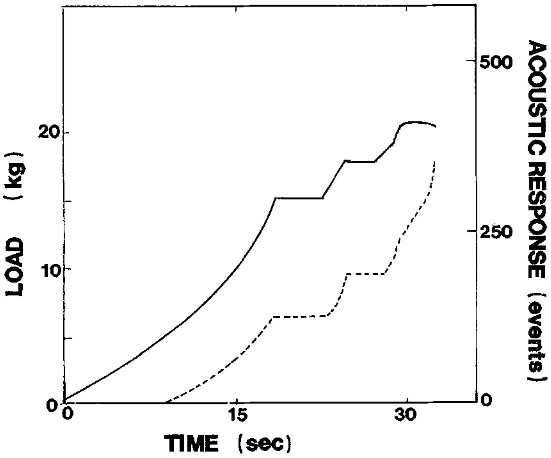

Figure 4 An alternative acoustic response method pre-cracking. The crosshead is withdrawn when emissions continue under a stationary load.

features in terms of the intermetallic primary particle size, the structure of the cobalt solid solution and the precipitate morphology.

The K_{IC} values for the "Tribaloys" are in the range 15 to 26 MN m$^{-3/2}$. These values are consistent with those reported for other hard metals, e.g. 20 MN m$^{-3/2}$ for WC–Co with 37% by volume fraction of the cobalt matrix [11]. For the two as-cast states and for most of the heat-treated conditions of both T400 and T800 the crack proceeded in a brittle manner through the matrix and the primary Laves phase. However, there was a preference for the crack path to favour the Laves particles through which it propagated in a cleavage mode (Figs. 5 and 6).

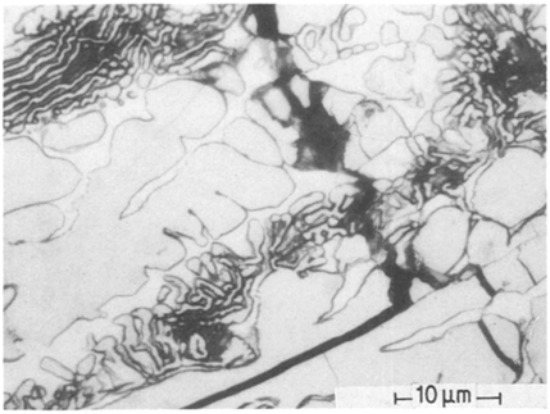

Figure 6 T400 as-cast microstructure showing the dominating effect of the large primary Laves particles in deviating the crack.

The brittleness of the transgranular fracture is reflected in the similarity in appearance of the fracture surface (Fig. 7) to that of an optical micrograph such as that shown in Fig. 1. The range of K_{IC} values for the microstructural conditions having this brittle transgranular fracture mode was 19.2 to 26.0 MN m$^{-3/2}$ for T400 and 18.7 to 24.2 MN m$^{-3/2}$ for T800. The highest K_{IC} for both "Tribaloys" corresponds to the 2 h at 1250° C heat-treated material and without these values the K_{IC} ranges are narrow, i.e. 19.2 to 21.9 MN m$^{-3/2}$ and 18.7 to 22.3 NM m$^{-3/2}$ for T400 and T800, respectively.

The two "Tribaloys" with a predominantly h c p solid solution and a Widmänstatten type precipitate had the lowest fracture toughness despite the fact that their hardness was identical to the as-cast alloys aged for 13 days at 800° C. The fracture mode in this instance was of a mixed intergranular and transgranular nature; the primary Laves phase no longer failed by cleavage but in an intergranular manner whereas the fracture path in the matrix

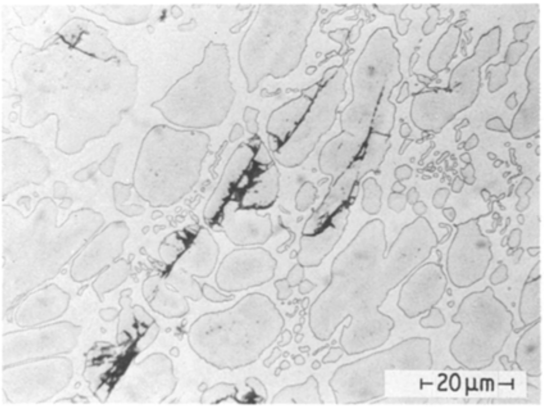

Figure 5 T800 as-cast microstructure showing early cleavage cracking in the primary Laves phase.

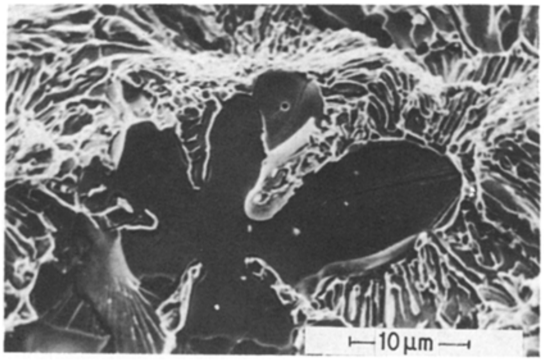

Figure 7 Typical scanning electron micrograph of the fracture surface of Tribaloy. Transgranular failure and the brittleness of the mode reflect the optical microstructure.

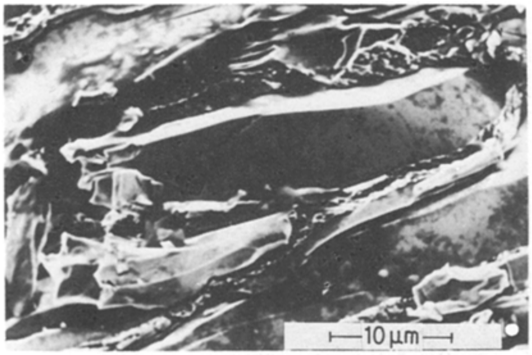

Figure 8 T800 aged 13 days at 800° C after 2 h at 1250° C. Scanning electron micrograph showing the intergranular fracture mode.

remained transgranular. Fig. 8 illustrates the intergranular nature of the failure.

More variation is seen (Table IV) in the MOR, than in K_{IC}. The highest value for a given Tribaloy was obtained with an f c c cobalt solid solution and no precipitate; this was also the alloy condition with the lowest hardness. The values of MOR generally decreased as the hardness increased for both T400 and T800. The exception to this trend being the very large decrease in the alloys with a predominantly h c p solid solution and Widmanstätten precipitate.

The calculated critical defect size enables an overall picture of the fracture behaviour to be assessed for T400 and T800. For both "Tribaloys" the largest values were associated with heat treatments that resulted in Widmanstätten precipitation in the h c p matrix.

4. Discussion

In most instances, the K_{IC} values of T400 are slightly higher than those of T800 in the same structural condition. This is because for most conditions the fracture is dominated by the cleavage of the primary Laves phase particles and T400 has a lower volume fraction of the intermetallic compound than T800. Although no information is available on the mechanical properties (other than room temperature hardness [5]) of the Laves phase in the "Tribaloys", Paufler and Schulze [12, 13] working on the deformation of single crystals of $MgZn_2$, found that completely brittle behaviour similar to that of a ceramic was exhibited up to about 450° C (~ 0.8 of the melting point T_m). Bilski [14] also found that polycrystalline Fe_2Nb, another Laves phase with the $MgZn_2$ type structure, did not soften until test temperatures in excess of 600° C (~ 0.5 T_m) . The Laves phase in "Tribaloys" melts at 1560° C which readily explains the brittle behaviour of this phase at room temperature and the fact that "Tribaloys" retain their wear resistance to high temperatures.

The only time that the fracture toughness behaviour of the "Tribaloys" is not dominated by the primary Laves phase is when the alloys are heat treated to give an h c p solid solution with a Widmanstätten precipitate. The matrix morphology leads to low K_{IC} values which are associated with a fixed failure mode, i.e. intergranular failure at the primary Laves phase/matrix interface and transgranular failure through the matrix. In these circumstances the matrix plays the major role in determining the low fracture toughness, hence T400 which has a greater volume fraction of matrix has the lower K_{IC}.

It is clear from the preceding discussion that except in the case of the h c p solid solution with Widmanstätten precipitation, the matrix structure has only a minor effect on the fracture toughness. This is not so for the MOR and, to explain the greater variation in these values, it is necessary to consider three distinct areas of microstructure, namely, the primary Laves, the lamellar eutectic and the areas of cobalt solid solution.

In the commercial as-cast state T800 has a higher volume fraction of the primary Laves phase, a much lower amount of the eutectic, but larger areas and a greater volume fraction of the Laves phase free areas of solid solution than T400 (see Table III). It is unlikely that the additional volume fraction of the brittle primary phase particles in T800 enhance the fracture behaviour, hence the higher MOR value is attributed to the presence of areas of solid solution.

In the commercial as-cast state, the solid components of the eutectic and the areas of solid solution are h c p and f c c. The substantial benefit of the f c c form of the cobalt solid solution is seen in both T800 and T400 after receiving the 2 h treatment at 1250° C when the highest modulus of rupture values were obtained for both "Tribaloys", the increase over that of the as-cast condition being most marked for T400. When the alloys are subjected to a rapid cool from the melt as in the arc-melted specimens the proportion of the solid solution with the f c c crystal is increased. In fact, for T400 only a trace of h c p is present and consequently the modulus of rupture is similar to that for the 1250° C treated sample. On the other hand, a significant proportion of h c p is still present in the arc-melted T800, there is a negligible amount of the intermetallic free solid solution and the volume fraction of primary Laves phase is increased to 70%. The detrimental effect of the increased amount of Laves and the absence of solid solution areas outweighs the gain due to the f c c solid solution and hence the modulus of rupture of arc-melted T800 is slightly less than that for the commercially as-cast state. The beneficial effect on the fracture behaviour on increasing the amount of f c c cobalt solid solution is consistent with the work of Bouguet and Dubois [15] on the tensile properties of cobalt. They found that the ducility increased as the amount of h c p was replaced by f c c.

The primary Laves phase cleaves easily and it is the varying ability of the solid solution component of the matrix to blunt the cracks by plastic deformation and so hinder the linking of microcracks that leads to the wide range of values for the modulus of rupture. The microhardness of the matrix of the alloys aged for 13 days at 800° C is harder than in the as-cast condition, indicating that plastic deformation is more difficult and this is reflected in the low modulus of rupture values compared to the as-cast condition. The lowest values of the MOR are associated with an h c p cobalt solid solution and Widmanstätten precipitation. In this situation the precipitate morphology is such that the matrix is more brittle than the primary Laves phase and as shown in Fig. 8, the fracture path is controlled by the matrix. The large flow size demonstrates the ease at which microcracking occurs through the matrix at stresses below those necessary to cleave the primary Laves phase.

5. Industrial significance

Components of "Tribaloys" are castable by investment and centrifugal methods; in addition the alloys can be used for hard-facing a softer base metal using oxyacetylene or gas-tungsten arc welding processes. It follows that the cooling rate from the melt will vary considerably with the production method. The present work has shown that the faster the cooling rate the greater the volume fraction of primary Laves phase and the larger the proportion of the f c c cobalt solid solution. Although these microstructural changes that depend on the cooling rate affect the room temperature mechanical properties as Table IV indicates, the effect of cooling rate on these properties is minimal. However, other properties in addition to the room temperature mechanical properties have to be considered, e.g. wear and elevated temperature behaviour. It is known that the f c c form of cobalt is less resistant to wear than the h c p [16], but any loss in wear resistance in rapidly cooled "Tribaloys" due to the predominance of an f c c solid solution is likely to be balanced by the gain associated with the increased proportion of hard Laves phase. At elevated service temperatures creep of the alloys may have to be considered and Kamel and Halim [17] have reported that the h c p crystal structure affords a greater resistance to creep than the f c c structure. However, more important than changes that might occur in the creep behaviour is the precipitation that takes place in the temperature range 600 to 1100° C. Spheroidal precipitates lead to a fall in MOR and increase critical defect size. This is undesirable, but tolerable particularly since there is a concomitant increase in hardness which may be beneficial for wear resisitance. Widmanstätten precipitate morphology which occurs on ageing the f c c solid solution, could be produced during service in alloys which have been rapidly cooled from the melt. This microstructural state gives the lowest K_{IC}, MOR and the largest flaw size and should be avoided.

After fabrication it is often desirable for the component to be stress relieved. If the component is to be used at elevated temperatures, then the cooling rate from the melt should not be rapid in order to allow the h c p solid solution to form; a component then stress relieved or put into service in the temperature range 600 to 1100° C would at least have a spheroidal precipitation. On the other hand, components to be used at room temperature and at moderate temperatures below 500° C, may be relieved at 1250° C which will produce an f c c solid solution. In this condition the alloys have the best room temperature fracture properties whilst still maintaining a reasonable hardness.

Finally, it is appropriate to consider alloy development. The excellent wear resistance of the "Tribaloys" is mainly due to the large volume fraction of Laves phase. It would appear that because the brittle intermetallic compound has to be present for the alloys to function in a wear-resistant capacity, the inherent resistance to crack propagation, as quantified by the critical stress intensity factor, will remain in the region of 20 $\mathrm{MN\,m^{-3/2}}$

In contrast, the present paper has shown that the resistance to crack initiation and growth can be altered significantly (see MOR and critical defect size data) and the benefit, in this context, of the fcc form of the solid solution has been demonstrated. Thus elements which stabilize the fcc crystal structure would be desirable additions to "Tribaloys". At the same time, in order to make full use of the excellent wear and corrosion resistance, at elevated temperatures, the alloying additions should also suppress the precipitation reactions, or at least prevent precipitation in the Widmanstätten morphology.

6. Conclusions

1. The resistance to brittle crack propagation (K_{IC}) for the "Tribaloys" lies in the range 15 to 26.0 $MN\,m^{-3/2}$ and because of the large volume fraction of the brittle Laves phase present, cannot be modified to any great extent.

2. For most conditions of the "Tribaloys" crack initiation and propagation takes place predominantly in a transgranular mode (cleavage of the primary Laves phase and transgranular through the matrix).

3. The "Tribaloys" resistance to crack initiation and growth to a critical size as measured by the modulus of rupture is affected significantly by the microstrucutre and can be varied between 304 to 923 $MN\,m^{-2}$.

4. The lowest fracture parameters were obtained when the cobalt solid solution was hcp with a Widmanstätten precipitate. In this condition the failure occurred in a mixed intergranular/transgranular manner (transgranular through the matrix and intergranular at the primary Laves particles) at stresses below those necessary to cleave the Laves phase.

5. The best room-temperature properties were associated with an fcc solid solution. This condition is only suitable for use at temperatures below 600° C, as in the range 600 to 1100° C precipitation in a Widmanstätten morphology would occur.

6. With the present compositions of the "Tribaloys" for service at elevated temperatures slow cooling from the melt is required. This will result in the equilibrium hcp solid solution which will give rise to the more favourable spherical precipitate morphology on exposure to high temperatures.

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References

1. Tribaloy wear resistant intermetallic materials (1979), Stellite Division Cabot Corporation Kokomo, Indiana.
2. C. B. CAMERON and D. P. FERRIS, *Cobalt* **3** (1974) 49.
3. R. D. SCHMIDT and D. P. FERRISS, *Wear* **32** (1975) 279.
4. A. HALSTEAD, PhD thesis University of London (1980).
5. A. HALSTEAD and R. D. RAWLINGS, *Met. Sci.* **18** (1984) 491.
6. R. R. WILLS, *J. Mater. Sci.* **11** (1979) 1330.
7. N. INGELSTON and H. NORBERG, *Eng. Fract. Mech.* **6** (1974) 97.
8. B. J. DALGLEISH, P. L. PRATT, R. D. RAWLINGS and A. FAKHR, *Mater. Sci. Eng.* **45** (1980) 9.
9. W. F. BROWN and J. E. STRAWLEY, ASTM STP No. 410 (1966).
10. G. K. BANSALL, *J. Amer. Ceram. Soc.* **59** (1976) 87.
11. S. B. LUYCKX, *Acta Metall.* **16** (1968) 535.
12. P. PAUFLER, *Phys. Status Solidi (b)* **52** (1972) 65.
13. P. PAUFLER and G. E. R. SCHULZE, *ibid* **24** (1967) 77.
14. A. BILSKI, MSc thesis, Imperial College, London University (1969).
15. G. BOUCHET and B. DUBOIS, *Scripta Metall.* **12** (1978) 1079.
16. D. M. BUCKLEY, *Cobalt* **38** (1968) 20.
17. R. KAMEL and K. HALIM, *Phys. Status Solidi* **15** (1966) 63.

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