Modeling of the Corrosion Behavior and Its Interrelation with the Deformation Behavior and Microstructure in a Newly Developed $\sim 7.5\text{Mn}-5\text{Cr}-3.0\text{Cu}$ Alloy White Iron

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Corrosion rate (CR), compressive strength (CS) and fracture strain (FS) were determined for different microstructures generated by heat-treating a newly designed $\sim 7.5\text{Mn}-5\text{Cr}-3.0\text{Cu}$ alloy white iron intended to be used for making castings to resist aqueous corrosion under marine conditions. The microstructures were characterized on the basis of size, shape, and distribution of the second phase, which in the present study comprised dispersed (DCs) and massive carbides (MCs). A new term, "distribution factor" (DF), has been evolved to characterize DCs. This factor proved useful in establishing a relationship between microstructure and CR, which can be represented by the expression

$$\text{CR} = [C_1 + C_2(\text{VMC}) + C_3(\text{VMC})^2] \cdot (\text{DF})^{C_4}$$

where C_1 , C_2 , C_3 , and C_4 are constants, VMC is the volume fraction of MCs, and DF is the distribution factor representing DCs. Corrosion rate was found to vary linearly with CS and FS in accordance with the equation

$$\text{CR} = (\text{CS} - A_1)/A_2 \quad \text{and} \quad \text{CR} = (A'_1 - \text{FS})/A'_2$$

where A_1 , A'_1 , A_2 and A'_2 are constants.

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I. Introduction

A comprehensive review of the literature on corrosion resistant alloy cast irons currently in use, namely, (a) ferritic (high Si), (b) austenitic (Ni-resist), and (c) martensitic (high Cr with or without Mo), revealed that the ferritic irons are most usefully employed under oxidizing conditions. Their poor mechanical strength and shock resistance preclude their general engineering applications⁽¹⁾⁽²⁾. Ni-resist irons, although very useful in a variety of environments, have a low strength and are unsuitable at operating temperatures $> 800^\circ\text{C}$ ⁽²⁾⁽³⁾. The high Cr irons can be employed upto higher service temperatures⁽²⁾. Their shock resistance is improved by lowering the carbon content⁽⁴⁾.

A critical analysis revealed that there is a general lack of systematic information on the electrochemical and on the deformation behaviour of microstructures encountered in alloy white irons. An alloy development program based on utilizing Mn, Cr, and Cu as alloying elements was therefore initiated by Patwardhan⁽⁵⁾ to assess whether new meaningful compositions, helpful in resisting aqueous corrosion, could be developed.

The basic idea involved is to develop corrosion resistant cast irons through the white iron route. It was anticipated that on doing so the shortcomings of high Ni-

cast irons (known as Ni-resist irons) namely their (i) low strength, (ii) being unsuitable for application at temperatures $> 800^\circ\text{C}$ and (iii) exhibiting graphitic corrosion and pitting will stand eliminated as graphite would be absent. To make the 'alloy development programme' additionally attractive it was felt appropriate to select the lower cost Fe-Mn-Cr-Cu system. The alloys were so 'designed' that (a) they would be cast white over a range of section sizes, (b) attained martensitic structure on air cooling/quenching from lower temperatures (800 and 850°C), and (c) could thus attain partly and completely austenitic matrices on increasing the heat treating temperatures up to 1050°C . This enabled a study of the electrochemical and deformation behavior of all the microstructures attainable in alloy white irons to be carried out with a view to optimizing 'composition' and 'microstructure'. The alloys thus developed were free from the aforesaid drawbacks observed in Ni resist irons, and attained corrosion rates not much inferior to those attained in Ni resist irons⁽²⁾⁽³⁾ even at this early stage of development and could thus be thought of as possible alternatives to them provided the programme is taken to its logical conclusion. A detailed account of the strategies adopted for the design of alloys and the experimentation carried out is reported else where⁽⁵⁾⁽⁶⁾. The alloys were investigated in detail in the as-cast and in the heat-treated conditions to arrive at qualitative interrelations between the microstructure and properties (corrosion and deformation behavior). Mathematical models inter-relating the corrosion behavior with (a) microstructure and (b) deforma-

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tion behavior proved useful in providing a quantitative assessment.

The aim of the present article is to highlight how the models interrelating (a) corrosion behavior with the microstructure and (b) corrosion behavior with the mechanical properties were developed.

II. Experimental Procedure

The alloy was air melted in a clay-bonded graphite crucible in a medium-frequency induction furnace and sand cast into 25-mm-diameter $\times$ 250 mm long cylindrical ingots and 8 mm $\times$ 22 mm $\times$ 120 mm rectangular strips. It was analyzed for C, S, P, and Si on a vacuum quantumeter and for Mn, Cr, P, and Si on an X-ray fluorescence spectrometer. The final composition is given in Table 1.

Hardness measurements and optical metallography were carried out on disc samples (14 to 18 mm in height) that were sliced off from the cylindrical ingots by making a 2- to 3-mm deep cut along the circumference on a silicon carbide cutoff wheel followed by tapping with a hammer. Heating of the specimens during cutting was minimized through water cooling. Specimens thus obtained were ground to have parallel faces. A similar procedure was adopted for preparing specimens for corrosion studies cut from rectangular strips.

Heat treatments primarily comprised soaking at 800, 850, 900, 950, 1000 and 1050°C for 2, 4, 6, 8 and 10 h, respectively, followed by oil quenching. They were carried out in muffle furnaces whose temperature was measured with a Pt-Pt/Rh (13 pct) thermocouple and controlled to $\pm 5^\circ\text{C}$.

Corrosion testing was carried out as per the relevant ASTM specifications⁽⁷⁾. After heat treatment of the rectangular specimens, a 1-mm thick layer was removed from all of the six faces by grinding. Each face was paper polished upto 4/0 to attain mirror finish. Specimens were then cleaned with double distilled water and finally with acetone, dried, weighed, and suspended for different durations in 100 cm³ capacity beakers containing 5% NaCl solution under stagnant conditions. After removal from the solution, specimens were cleaned, dried, and weighed⁽⁷⁾. From the mass loss, the corrosion rates (CR in kg/m²·s) were calculated by using the formula⁽⁸⁾.

$$\text{CR} = \frac{K \cdot W}{A \cdot T}$$

where

$$K = 2.78 \times 10^{-3}$$

W = mass loss in g nearest to 1 mg

A = surface area in cm² nearest to 0.01 cm²

T = duration in h nearest to 0.01 h

Table 1 Chemical analysis of the alloy (mass%).

C	Si	S	P	Mn	Cr	Cu
2.85	1.80	0.072	0.305	7.3	4.5	2.86

Hardness measurements were carried out on a Vickers hardness testing machine employing a 29.4 N load. A total of 20 impressions were recorded on both the faces of a specimen. The permissible scatter in the hardness values was ± 17 VPN⁽⁹⁾.

Specimens of the size 10 mm in diameter $\times$ 10 mm in height were prepared from cylindrical ingots for deformation studies comprising compression testing. Tests were carried out on the as-cast and on the heat-treated specimens on a 588 kN capacity MFL microprocessor based universal testing machine at a crosshead speed of 1.0 mm/min. Compressive strength (CS) and the percent deformation were calculated from the load deformation curves in the usual manner.

Quantitative metallographic studies were carried out on a LEITZ image analyzer at magnification 2500 times. Specimen size was the same as that employed during hardness measurements. Quantitative estimations and data processing, including plotting of histograms, etc., were carried out with the help of computational techniques on a DEC-2050 and PC/AT systems.

III. Results and Discussion

The proposed investigation was undertaken to evolve models interrelating corrosion behavior with the microstructure with a view to optimize it, especially with regard to the size, shape, and distribution of the second phase. This calls for generating different microstructures, characterizing them on the basis of their corrosion and deformation behavior, and finally selecting an appropriate microstructural base for further analysis.

It would be pertinent to mention here the relevance of physical metallurgical parameters in controlling corrosion. A single-phase microstructure (preferably solid solution) with a close-packed crystal structure is most useful in resisting corrosion. Presence of a passive film would be an added advantage. Their main drawback is relatively lower strength. For achieving an optimum combination of corrosion resistance and mechanical properties, the hard second phase should have a spherical/polygonal morphology. Their volume fraction should neither be too large (leading to enhanced galvanic action) nor too small (promoting pitting). Metastable microstructures, such as martensite, also lead to smaller CR due to high hardness. However, a combination of a hard second-phase dispersion in a martensitic matrix would be extremely brittle and therefore not conducive to attaining a useful combination of high strength and toughness. Electrochemically, the difference in potential between the matrix and the second phase should be as small as possible. With specific reference to alloyed cast irons, the main interest would center around martensite and austenite based structures with second phase as massive (MCs) and dispersed carbides (DCs)⁽⁵⁾⁽⁶⁾ for reasons stated above. Both the eutectic and the proeutectoid carbides constitute MCs. DCs directly formed from austenite during soaking irrespective of the matrix. In the as-cast state and on heat treating from upto 900°C the

carbide is M_3C type; boundary carbides $M_{23}C_6$ upto 900°C 10 h heat treatment may also persist upto 950°C 2 h heat treatment. On heat treating from 950 and 1000°C M_5C_2 and M_7C_3 formed whereas M_7C_3 was the only carbide to form at 1050°C .

The results and discussion are therefore put forth in three stages dealing with the (1) effect of heat treatment on structural changes and hardness, (2) evolution of a basis for quantifying second phase in the form of DCs and (3) development of the models, first interrelating the corrosion behaviour with the microstructure and, second, interrelating deformation behaviour with the corrosion behaviour.

1. Effect of heat treatment on the hardness and microstructure

The heat-treating experiments were primarily designed to assess the feasibility of attaining different microstructures of interest and to characterize them initially on the basis of hardness. Oil quenching was preferred over air cooling to minimize precipitation of carbides during cooling. This would ensure that their possible extraneous effect on properties is minimized.

The effects of time and temperature on the hardness are summarized in Fig. 1. The data points contained in the figure represent experimentally determined values, whereas the actual plots correspond to the best-fit data. A detailed discussion on the interrelation between hardness and the microstructure as influenced by heat-treating temperature and time has been reported elsewhere⁽⁶⁾ and is being excluded since it does not fall within the scope of the article. However, its summary, relevant to the theme under discussion, has been included.

The as-cast microstructure comprised martensite/pearlite+carbide+some retained γ . Its hardness is marked by the point 'X' on the ordinate of Fig. 1. The continuous although a small increase in hardness with respect to the as-cast state, on heat treating from 800°C , is due to an increase in the volume fraction of martensite

(Fig. 1). This can be explained on the basis of the reaction

The volume fraction of martensite increases with time as more DCs form out of γ .

On heat treating from 850°C , the initial decrease in hardness at 2 h soaking period is due to a further increase in the γ -stabilizing tendency. This reduces the volume fraction of martensite in the final microstructure. The temperature is too low to influence MCs. Therefore, the overall hardness is independent of the soaking period because apart from the initial change that occurred on heat treating at 2 h, increasing the soaking period further has no further tangible effect on the microstructure. The lower overall hardness compared to that attained at 800°C is due to the retention of austenite.

Increasing the soaking period (2 to 10 h) has little effect on the hardness on heat treating at 900°C , as the microstructure is practically unaltered except for a limited coarsening of the DCs and some reduction in the volume fraction of MCs (Fig. 1, Table 2). The lower overall hardness at 900°C results because the matrix is by and large austenitic.

The structural changes at 950°C are similar to those occurring at 900°C except that the coarsening of DCs is further enhanced and reduction in the volume fraction of MCs is larger (Table 2) due to the availability of a larger activation in terms of temperature (Fig. 1). These changes which promote the retention of a relatively larger volume fraction of stable austenite not only result in a lower overall hardness compared with that observed on heat treating at 900°C but appear to be confined to the first 2 h. This explains the reasons for hardness to be independent of time.

The hardness values on heat treating at 1000 and 1050°C can be similarly explained, the decrease in volume fraction of MCs being marked, which would lead to a steep fall in hardness with soaking period.

2. Effect of heat treatment on the volume fraction and morphology of massive carbides

Massive carbides in the as-cast state are only partly

Fig. 1 Effect of soaking period on hardness.

Table 2 Effect of heat treatment on the volume fraction of massive carbide.

Temperature (°C)	Volume fraction (%)*			
	2 h	4 h	6 h	10 h
800	19.2	—	—	18.9
850	20.3	20.2	19.6	18.0
900	21.2	19.0	17.9	16.4
950	21.0	20.6	21.1	18.5
1000	18.6	—	—	15.0

*Average of 10 observations

discontinuous. This is consistent with the hypoeutectic character of the alloy. On heat-treating, their volume fraction has been reduced, and they are rendered discontinuous due to the combined graphitizing action of copper and silicon. This tendency increases with heat-treating temperature and time and is expected to be dominant at temperatures $>950^{\circ}\text{C}$, consistent with physical metallurgical considerations associated with malleabilizing⁽¹⁰⁾. The proposed analysis explains the data summarized in Table 3. Other reasons why volume fraction (Vf) of MC may not significantly decrease until 950°C are (1) other transformations requiring lesser activation *e.g.* formation of martensite and DCs take precedence and (2) fully austenitic matrices favour transformations involving MC (*e.g.* such as reducing their volume fraction) and this unequivocally occurs only $\approx 950^{\circ}\text{C}$ ⁽⁶⁾.

An equally important aspect needing consideration is that unlike in malleable irons, the carbide phase has been rendered stable by Cr additions, and the presence of Mn renders it additionally stable as nearly 50 to 55 pct of the Mn added also partitions to it⁽¹¹⁾. Therefore, as the heat-treating temperature and time are increased, the MCs, rather than decomposing into graphite, would acquire a low-energy configuration/morphology, such as near spherical or hexagonal. This analysis explains the general

'rounding off' observed in MCs on heat-treating from higher temperature, *i.e.*, $>950^{\circ}\text{C}$ ⁽⁶⁾. Quite evidently no graphite forms.

3. Effect of time and temperature on the size and distribution of dispersed carbides

The results summarized in Tables 3 and 4 are helpful in characterizing DCs fully. As can be seen, particles constituting the DCs initially belong to classes I (covering a size range upto $0.58\text{ }\mu\text{m}$ and II (covering a size range from 0.58 to $1.16\text{ }\mu\text{m}$ because they exclusively fall into these two classes at the formation stage. On heat-treating, their distribution is altered in a manner consistent with the attributes of a nucleation and growth type of transformation, leading to coarsening. This would involve (i) a reduction in the number of particles in the first two classes and simultaneous increase in the number of particles in classes III and IV and (ii) an increase in the mean diameter (Tables 3 and 4).

A critical analysis of the data (Tables 3 and 4) reveals that it would be difficult to arrive at an interrelation between the heat-treating parameters and the extent of coarsening based on the Ostwald's ripening concept⁽¹²⁾. Of greater significance would be to quantify the distribution of the second phase particles as influenced by heat treatment, as this, together with the extent of coarsening, is likely to influence properties in general and corrosion behavior in particular. This difficulty was overcome by evolving a new parameter called the "distribution factor" (DF) defined as

$$\text{DF} = \frac{\sum_{i=1}^n N_i \cdot X_i}{\sum_{i=1}^n N_i} \quad (1)$$

where

n = the number of classes

N_i = the number of particles in the i th class ($i = 1, 2, 3, \dots, n$)

X_i = volume fraction in the i th class/VDC

VDC = total volume fraction of dispersed carbides.

The DF was calculated on the basis of particle distribution at each of the 10 fields of observation for a given heat treatment and then averaged (Table 5).

Qualitatively speaking, the smaller the DF is, either individually or in comparison with a reference base, the larger would be the extent of coarsening. Thus, the DF can be usefully employed to quantify the extent of coarsening⁽⁶⁾ without the necessity of employing the Ostwald's equation⁽¹²⁾,

$$r_1^3 - r_o^3 = k(t_1 - t_o)$$

where

r_1 = particle radius at time t_1 , and

r_o = particle radius at time t_o .

This only takes into consideration the average particle size and does not in any way reflect upon the changes in

Table 3

(a) Effect of heat treatment on the mean diameter of dispersed carbides.

Temperature ($^{\circ}\text{C}$)	Mean diameter μm^*			
	2 h	4 h	6 h	10 h
800	—	—	—	0.47
850	0.38	—	—	0.51
900	0.58	0.53	—	0.70
950	0.57	0.60	—	0.77

(b) Effect of heat treatment on the average number of particles.

Temperature ($^{\circ}\text{C}$)	Number*			
	2 h	4 h	6 h	10 h
800	—	—	—	66
850	55	—	—	49
900	42	46	—	36
950	47	37	—	33

(c) Effect of heat treatment on the volume fraction of dispersed carbides.

Temperature ($^{\circ}\text{C}$)	Volume fraction (%)*			
	2 h	4 h	6 h	10 h
800	—	—	—	12.2
850	7.4	—	—	9.1
900	9.9	9.2	—	10.3
950	10.3	8.8	—	12.2

*Frame area = $359.75\text{ }\mu\text{m}^2$; each reading represents an average of 10 values.

Table 4 Effect of heat treatment on the percent (a) number and (b) area of dispersed carbides in different classes.

(a)																
Temperature (°C)	Class I (h)				Class II (h)				Class III (h)				Class IV (h)			
	2	4	6	10	2	4	6	10	2	4	6	10	2	4	6	10
800	—	—	—	70	—	—	—	30	—	—	—	1	—	—	—	0
850	82	—	—	61	17	—	—	37	0	—	—	2	0	—	—	0
900	53	60	—	41	40	37	—	45	6	3	—	13	0	0	—	1
950	50	51	—	46	42	42	—	31	8	7	—	16	0	0	—	7

(b)																
Temperature (°C)	Class I (μm)				Class II (μm)				Class III (μm)				Class IV (μm)			
	2	4	6	10	2	4	6	10	2	4	6	10	2	4	6	10
800	—	—	—	8.5	—	—	—	3.6	—	—	—	0.08	—	—	—	NIL
850	6.1	—	—	5.6	1.3	—	—	3.4	0.02	—	—	0.15	NIL	—	—	NIL
900	5.3	5.5	—	4.3	4.0	3.4	—	4.7	0.62	0.31	—	0.32	NIL	NIL	—	0.08
950	5.8	4.5	—	5.6	3.8	3.7	—	3.8	0.67	0.60	—	1.92	NIL	0.03	—	0.86

Class I=upto 0.58 μm; Class II=0.58 to 1.16 μm; Class III=1.16 to 1.74 μm; Class IV=1.74 to 2.32 μm;

Class V=2.32 to 2.90 μm; Class VI=2.90 to 3.48 μm.

Frame area=359.75 μm²; each reading represents an average of 10 values; class width=0.578 μm (=0.58 μm)

Table 5 Effect of heat treatment on the distribution factor.

Temperature (°C)	Soaking period (h)	Distribution (DF)
800	10	0.582
850	2	0.736
	6	—
	10	0.534
900	2	0.479
	4	0.519
	6	—
	10	0.410
950	2	0.483
	4	0.461
	6	0.440
	10	0.359

Table 6 Effect of heat treatment on the corrosion behavior.

Hardness Hv30	Condition	Corrosion rate** (10 ⁻⁹ kg·m ⁻² ·s ⁻¹)
621	As cast	36.52
476	900°C 4 h, OQ*	29.84
476	900°C 10 h, OQ	26.82
441	950°C 4 h, OQ	26.97
444	950°C 10 h, OQ	30.17

*OQ=oil quenched

**1 × 10⁻⁹ kg·m⁻²·s⁻¹=0.864 mg·dm⁻²·d⁻¹ (mdd)

the distribution brought about by heat-treating.

4. Mathematical modeling of the corrosion behavior

The data summarized in Table 6 reveal that the DCs and MCs have an important bearing on the corrosion behavior of the alloy with an austenitic matrix, i.e., when heat treated from 900 and 950°C. It was therefore considered appropriate to develop the proposed model incorporating the possible effects of the γ-matrix, MC, and DC. Accordingly, the CR can be expressed as

$$CR = F(vf/stability, vf \text{ of MC}, vf \text{ of DC}, \text{distribution of the DC}). \quad (2)$$

The parameter $f(\gamma \text{ vf/stability})$ essentially signifies volume fraction and stability of γ. The larger the volume fraction of γ the better would be the corrosion resistance. For a given volume fraction of γ, the higher its stability, the better would be the corrosion resistance.

Excluding the last term to start with, and as a first approximation, the contribution of the volume fraction and the stability of γ may be treated as nearly a constant,

because the hardness of the alloy heat-treated at 900 and 950°C is approximately the same; i.e., changes in the properties are being essentially controlled by changes in MC and DC.

The variation in CR with the volume fraction of MCs revealed that the functional relationship can be represented by a second-order polynomial:

$$CR = A + B(VMC) + C(VMC)^2 \quad (3)$$

where VMC=volume fraction of MCs.

The constants A , B , and C were calculated by the least square method⁽¹³⁾⁽¹⁴⁾. After incorporating the DF, the above equation assumes the form

$$CR = [C_1 + C_2(VMC) + C_3(VMC)^2] \cdot (DF)^{C_4} \quad (4)$$

The constants C_1 , C_2 , C_3 , and C_4 were calculated and computed to minimum error by using a nonlinear multivariable constraint optimization technique⁽¹³⁾⁻⁽¹⁵⁾ and the final expression is

$$CR [\text{kg} \cdot \text{m}^{-2} \cdot \text{s}^{-1}] = 10^9 \cdot [-216.2 + 26.528(VMC) - 0.716(VMC)^2] \cdot (DF)^{-0.02} \quad (5)$$

The theoretically calculated values of CR based on the above model are in good agreement with the experimen-

Fig. 2 Calculated vs observed corrosion rates.

tal data (Fig. 2).

5. Interrelation between deformation and corrosion behavior

Since one of the principal objectives of the present investigation is to examine whether an interrelation existed, the relevant data pertaining to compressive properties and CRs were collated (Tables 6 and 7). It is observed that as CS and fracture strain increase, CR decreases. Recalling that the matrix is austenitic, this observation appears logical, since the parameter(s) leading to an increase in brittleness (*e.g.*, the unfavorable distribution of coarsened carbides) would also enhance CR.

Through an analysis of relevant data, it emerged that the relation between CS and CR and that between fracture strain and CR can be represented by an expression

$$X = C_1 + C_2(Y) \quad (6)$$

where

Y = property (CS in MPa or fracture strain in %)

X = CR in $\text{kg} \cdot \text{m}^{-2} \text{s}^{-1}$

C_1 and C_2 = constants.

Table 7 Effect of heat treatment on the deformation behavior.

Condition	CS (MPa)	Fracture strain (%)
As cast	2352	12.0
900°C 4 h, OQ	2219	27.3
900°C 10 h, OQ	2288	28.3
950°C 4 h, OQ	2341	28.1
950°C 10 h, OQ	2297	28.4
1000°C 4 h, OQ	2683	30.9
1000°C 10 h, OQ	2845	31.8

These constants were evaluated using the least-squares method, and final equations are

$$\text{CS} = 2154 + 0.138 \times 10^9 (\text{CR})$$

or

$$\text{CR} [\text{kg} \cdot \text{m}^{-2} \text{s}^{-1}] = \frac{\text{CS} [\text{MPa}] - 2154}{0.138 \times 10^9} \quad (7)$$

$$\text{fracture strain} [\%] = 54.3 - 0.91 \times 10^9 (\text{CR})$$

or

$$\text{CR} [\text{kg} \cdot \text{m}^{-2} \text{s}^{-1}] = \frac{54.3 - \text{fracture strain} [\%]}{0.91 \times 10^9} \quad (8)$$

Measured values of CR agree well with the calculated values of CR (based on CS), as revealed by Figs. 3 and 4.

Before concluding it would be appropriate to arrive at some further rationalizations. On interrelating eqs. (5), (7) and (8), it emerges that

$$\text{CS} [\text{MPa}] = 2154 + 0.138 [-216.2 + 26.53 (\text{VMC}) - 0.716 (\text{VMC})^2] \cdot (\text{DF})^{-0.02} \quad (9)$$

and

$$\text{CS} = 2162.7 - 0.15 (\text{fracture strain} [\%]) \quad (10)$$

Like in eq. (9), the value of % strain can be expressed in terms of the microstructure by solving eqs. (9) and (10) which yields that

Fig. 3 Experimental vs predicted CR based on compressive strength.

Fig. 4 Experimental vs predicted CR based on fracture strain [%].

$$\text{fracture strain [\%]} = 56.13 - 0.921[216.2 + 26.53(\text{VMC}) - 0.716(\text{VMC})^2] \cdot (\text{DF})^{-0.02} \quad (11)$$

A critical review of the eqs. (9) and (11) indicates that whereas CS is mostly governed by the matrix microstructure and not a strong function of MCs and distribution of DCs, % strain is influenced to a much larger extent by the volume fraction of MCs and distribution of DCs. This is a reasonable appraisal looking to the nature of the material/microstructure under review.

IV. Conclusions

(1) A model interrelating CR with microstructure is successfully developed.

(2) In this model, the dispersed second-phase particles are, perhaps for the first time, characterized on the basis of a newly evolved parameter termed the distribution factor.

(3) Deformation behavior, represented by CS and fracture strain, is linearly related with CR.

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