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Effect of C content on the mechanical properties of solution treated as-cast ASTM F-75 alloys

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The mechanical properties of solution treated ASTM F-75 alloys with various carbon contents have been studied. Alloys cast under the same conditions were subjected to solution treatment for several periods and then their tensile properties were evaluated. In the as-cast conditions, the alloys exhibited higher strength values with increasing carbon content whereas their ductility was not significantly affected. For the solution treated alloys, the variation of the strength was characterized by a progressive increase for short treatment times until a maximum value was achieved, which was followed by a diminution in this property for longer treatment times. This behavior was more accentuated for the case of the alloys with medium carbon contents, which also exhibited the highest values of strength. Furthermore, the alloy's ductility was enhanced progressively with increasing solution treatment time. This improvement in ductility was significantly higher for the medium carbon alloys compared with the rest of the studied alloys. Thus, high and low carbon contents in solution treated ASTM F-75 alloys did not produced sufficiently high tensile properties.

1. Introduction

The Co-Cr-Mo-C alloy complying with the ASTM F-75 standard is used in surgery implants [1, 2]. The components manufactured by investment casting exhibit a microstructure consisting of a cored Co-rich face centered cubic (fcc) matrix with interdendritic and grain boundary precipitates [1, 3]. The precipitation of carbides represents the major strengthening mechanism in the as-cast conditions, and it is also mostly responsible for the low ductility observed in the alloy under such conditions, which is frequently insufficient to satisfy the requirements of the aforementioned standard [2–5]. Thermal processing constitutes the most commonly used alternative to improve the alloy's ductility by modifying the amount and size of the carbide precipitates.

The ASTM F-75 alloys exhibit rapid work hardening, which is commonly attributed to the intrinsic stacking faults [6, 7]. The low stacking fault energy associated with the matrix of these alloys difficults the motion of extended dislocations. Additionally, under the effect of external stresses the creation of stacking faults is promoted by the low stacking fault energy associated with an increased plastic deformation. The high density of the stacking faults developed in this way restricts the motion of dislocations, resulting in a strengthening of the alloy. Rajan and Van der Sande [7] considered the latter mechanisms to be the main contributors to the alloy's work hardening. On the other hand, Kilner *et al.* [8] associated the initial alloy's work hardening and yield strength to the presence in it of coarse carbide precipitates. The observations of the latter authors showed that the alloy's work hardening increased with its carbide content. However, these observations arouse from the analysis of alloys thermally treated under different conditions; consequently, no definitive conclusions can be drawn from their results with respect to the contribution of carbide precipitates to the observed work hardening.

Although the literature reports increased mechanical properties as a result of a solution treatment given to the as-cast alloy, a wide variation exists in the values of the resulting properties reported as well as in the variables involved in the solution treatments employed [6, 9–13]. It is accepted that the final microstructure, which is responsible for the resulting properties, depends on variables such as chemical composition and as-cast microstructure of the alloy, as well as time and temperature of the solution treatment. Regarding the chemical composition of the alloy, the carbon content level influences the as-cast microstructure, greatly determining the distribution, amount, size and type of the carbide precipitates present in the alloy [10, 12]. On the other hand, the carbide precipitates evolve during

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the progress of the solution treatment so that the final microstructure and therefore the resulting mechanical properties depend on the treatment time employed.

It is clear that a number of parameters influence the microstructure of solution treated alloys and that a proper design of the thermal treatment is required in order to achieve appropriate mechanical properties. Furthermore, the contribution of carbide precipitates to the mechanical properties of these alloys has not been entirely rationalized, so that further work is needed in order to clarify this point. Thus, the aim of this work was to study the effect of solution treatment on the mechanical properties of as-cast ASTM F-75 alloys containing various carbon contents.

2. Experimental procedure

Three Co-Cr-Mo-C alloys with variable carbon content, all of them complying with the ASTM F-75 standard, were employed. Two runs were conducted for each alloy using an induction furnace, in which the alloys were heated, melted and cast at 1873 K into a ceramic mold pre-heated at 1223 K, under an argon atmosphere. The chemical analysis of the resulting alloys is given in Table I, in which the requirements of the ASTM F-75 standard are also included for comparison. Table I shows that for the cast alloys the resulting carbon contents were 0.16, 0.23 and 0.31 wt%. According to this, the alloys were labeled as "low C", "medium C" and "high C" alloy, respectively.

The mold included a set of 11 cylindrical cavities, each one with a length of 10 cm and a diameter of 1.5 cm. The mold was designed in such a way as to allow all cavities to have the same filling and cooling conditions in order to minimize any possible microstructural variations among the solidified bars due to these causes.

The solution treatment temperature should not exceed the melting point of the interdendritic material. Thus, in order to obtain an estimation of the latter temperature, some specimens of each alloy were heated and held for 240 min at different temperatures and then quenched in water. The tested temperatures were progressively increased by 10 K steps, starting at 1473 K. The melting of carbides, detected under an optical microscope, was found to occur at 1513 K and higher temperatures but it was not observed at temperatures below 1503 K. Based on these observations, 1503 K was selected as the temperature to use in subsequent solution treatments and carbide dissolution studies. The temperature interval in which the melting of the carbide precipitates occurred, 1503–1513 K, agrees relatively well with the final temperature of solidification, according to the literature reports [10, 15, 16].

In order to follow the variation of the carbide content as a function of treatment time, a carbide precipitate dissolution study was conducted for each alloy. These studies were carried out at the pre-determined solution treatment temperature of 1503 K. Specimens with a length of 6 mm were cut off from one of the cast bars, which was previously machined to a final diameter of 5 mm in order to remove the chill shell. The machined probes were encapsulated inside evacuated quartz tubes to prevent their oxidation during the solution treatments. Then, the specimens were heated up to 1503 K and held at this temperature for intervals of time ranging from 0 to 240 min, followed by quenching in water. Specimens heated up to 1503 K and then water quenched, with no holding time at that temperature, corresponded to the samples with "zero treatment time". Optical microscopy coupled with image analysis was used for the microstructural characterization of the specimens, which were previously conventionally prepared by metallographic grinding and polishing. Each quantitative determination of the carbide content was made at 100× on 15 fields, which accounted for a total analyzed area of about 1.9 mm^2 per specimen.

The solution treatments given to the as-cast bars were conducted in an electric resistance chamber furnace. The thermal cycle employed included a rapid heating up to the predetermined solution treatment temperature and then holding at that temperature for intervals of time of 0, 30, 60, 90, 120 and 240 min, following this by quenching in water.

The tensile properties, namely ultimate tensile strength (UTS), yield strength (YS) and elongation to fracture (%), were evaluated on the as-cast and solution treated specimens; at least two replicas were tested per each set of solution treatment conditions. The tested specimens were machined from the cast bars according to the ASTM E-8 standard and the tension tests were conducted using a 15 ton INSTRON machine, with a deformation rate of 3 mm min^{-1}.

TABLE I Chemical composition of the cast alloys

	Element (wt%)							
ASTM F-75 alloys	Cr	Mo	C	Si	Mn	Ni	Fe	Co
Nominal composition	27–30	5–7	0.35 max	1 max	1 max	1 max	0.75 max	Bal.
Low C								
Run 1	27.03	5.8	0.17	0.68	0.47	0.70	0.74	Bal.
Run 2	27.01	5.6	0.17	0.66	0.46	0.72	0.75	Bal.
Medium C								
Run 1	27.06	6.0	0.23	0.94	0.36	0.73	0.49	Bal.
Run 2	27.04	6.0	0.23	0.93	0.37	0.72	0.48	Bal.
High C								
Run 1	27.00	5.8	0.30	0.60	0.38	0.57	0.53	Bal.
Run 2	27.02	5.8	0.30	0.60	0.37	0.58	0.51	Bal.

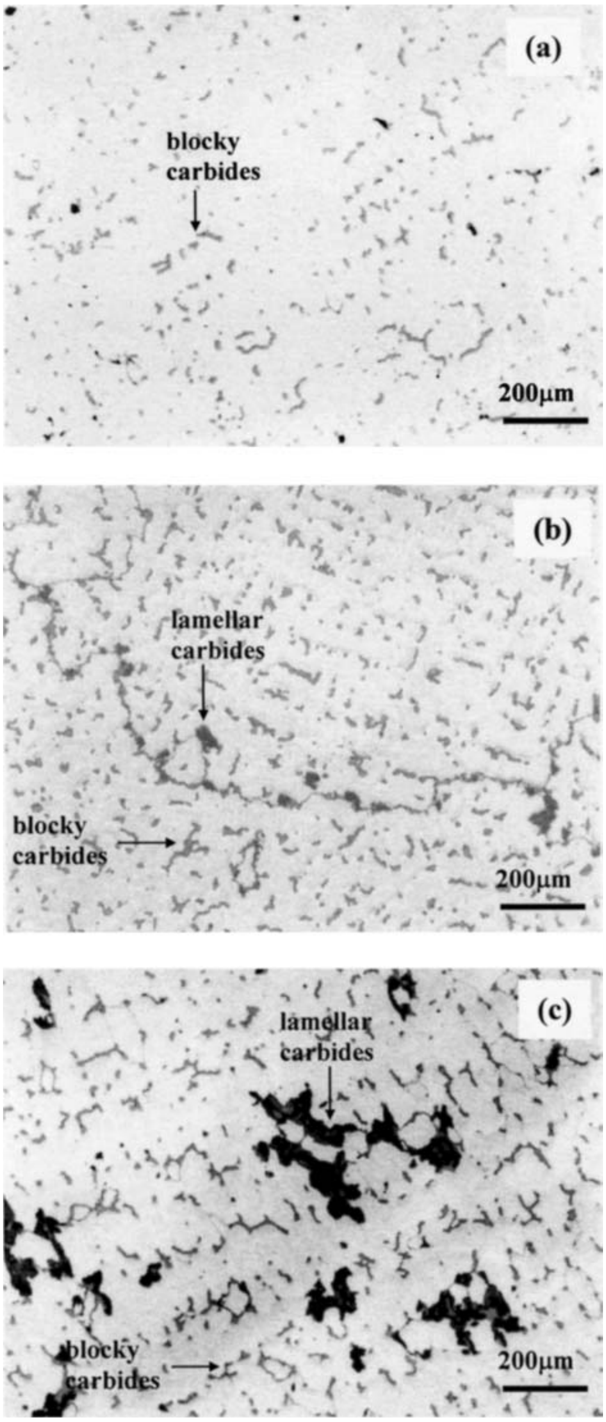

Figure 1 As-cast microstructures of the studied alloys: (a) low, (b) medium, and (c) high C alloys.

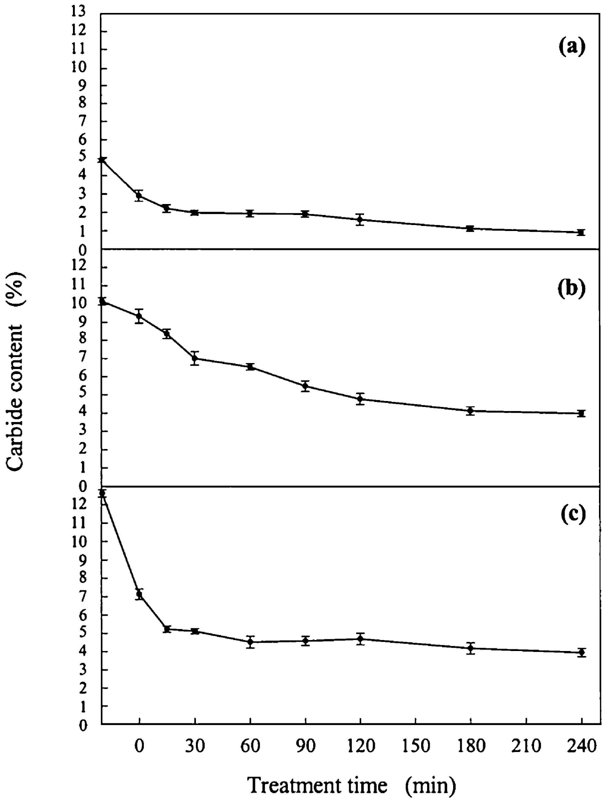

Figure 2 Evolution of carbide content as a function of treatment time: (a) low, (b) medium, and (c) high C alloys.

3. Results and discussion

Fig. 1 shows typical as-cast microstructures observed in the studied alloys. In the low C alloy, Fig. l(a), "blocky" precipitates were observed in the interdendritic regions. An increased carbon content gave rise to a bigger amount of blocky precipitates, promoting also the formation of lamellar precipitates at the grain boundaries and immediate areas, Figs. 1(b) and (c). It is thought that in the medium and high C alloys, the solidification conditions promoted a carbon segregation which led to the precipitation of lamellar carbides during cooling of the solidified alloys.

The variation of the carbide content with increasing solution treatment time at 1503 K is shown in Fig. 2. In this figure the limits of the error bars correspond to the 95% confidence interval. For all three alloys, the carbide content decreased markedly for the relatively short treatments. Furthermore, for treatment times varying between 120 and 240 min, the alloy's carbide content showed only a slight variation. Hence, the treatment times chosen in order to follow the evolution of the mechanical properties of the alloys as a function of their carbide content were 0, 30, 60, 90, 120 and 240 min. Notice in Fig. 2 that for the first 90 min of solution treatment, the carbide content in the medium C alloy was higher than that in the high C alloy, although the initial carbide content, i.e., in the as-cast conditions, was lower for the former alloy. This is probably due to the higher content of lamellar carbides in the as-cast conditions for the high C alloy as well as to a faster dissolution of this type of carbide [16]. Hence, once lamellar carbides dissolved, the content of the remaining blocky carbides was lower in the high C alloy.

Figs. 3–5 show the levels of the tensile properties evaluated in the as-cast as well as in the solution treated alloys. The values specified by the ASTM F-75 standard are indicated by discontinuous lines. Fig. 3 illustrates the variation of YS with solution treatment time. As can be seen, the duration of the solution treatment did not influence significantly the YS levels in the low and high C alloys, both of which exhibited YS values close to that established by the ASTM F-75 standard. In contrast, the YS values for the medium C alloy were superior to that specified by the standard and were characterized by an increment with increasing treatment time until a maximum value was

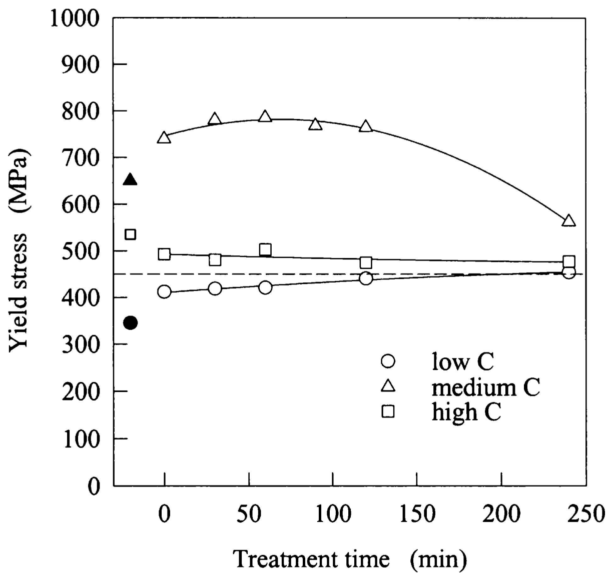

Figure 3 Effect of treatment time on yield stress of as-cast ASTM F-75 alloys. Filled symbols denote the as-cast conditions.

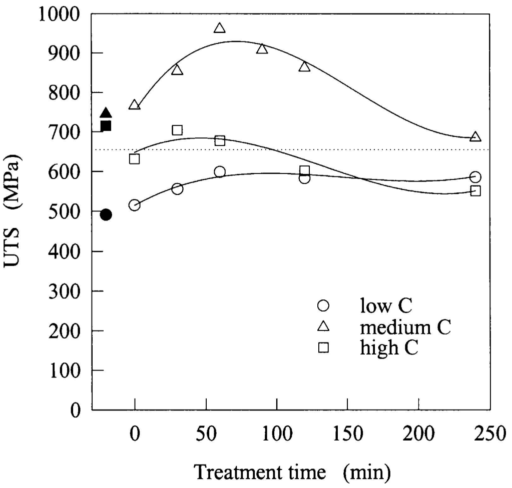

Figure 4 Effect of treatment time on UTS of as-cast ASTM F-75 alloys. Filled symbols denote the as-cast conditions.

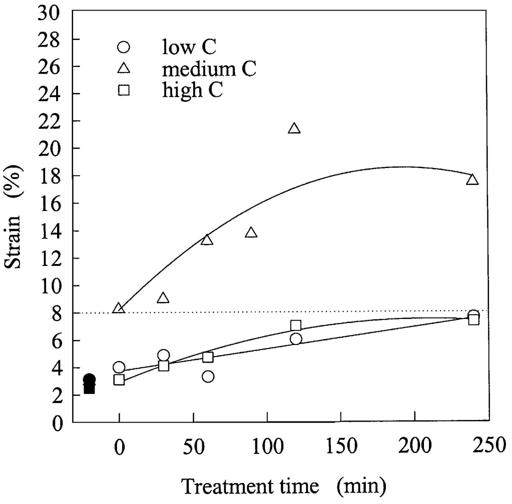

Figure 5 Effect of treatment time on strain of as-cast ASTM F-75 alloys. Filled symbols denote the as-cast conditions.

reached, which was followed by a posterior continuous diminution.

The effect of the solution treatment time on the alloys' UTS is shown in Fig. 4. For short treatment times, regardless of the alloy, the UTS exhibited a behavior characterized by a progressive increase. This was particularly significant for the case of the medium C alloy, whereas for the low and high C alloys the increment was less significant, being approximately of the same order of magnitude for both alloys. On the other hand, for long solution treatment times the UTS resulted to be practically unaltered for the case of the low C alloy, whereas it was progressively decreased in the medium and high C alloys. It is worth noting that for the high C alloy the progressive diminution of the UTS led to values located below the limit established by the standard. Regarding the ductility behavior, the solution treatment increased ductility in all treated alloys, as illustrated in Fig. 5. However, whereas the values corresponding to the treated medium C alloy exceeded the standard value, those for the low and high C alloys exhibited insufficient levels, despite their progressive improvement with increasing solution treatment time.

In the as-cast conditions, the low C alloy showed the lowest UTS value as a result of its low carbide content. However, for the medium and high C alloys the UTS levels were comparable, despite the significant difference in the total carbide content existing between both alloys. This result was associated to a brittle effect induced by the lamellar carbides present in these alloys. This result agrees well with the observations of Montero *et al.* [12] in as-cast alloys with 0.29 and 0.35% C. The ductility obtained in the as-cast conditions was similar for all three alloys, about 3%, indicating that the effect of the carbide precipitates on this property is independent of their type, amount and distribution.

The solution treatment induced significant changes in the as-cast microstructures. For short treatment times, the microstructural evolution was characterized by a rapid decrease in the total carbide content, Fig. 2. This in turn promoted an increase in the alloys' strength and ductility. It is noteworthy the enhanced ductility in the medium C alloy. The enhanced strength attributed to the diminution of carbide content agrees well with the observations of Dobbs and Robertson [9] for a 0.25% C alloy, however, it contradicts the findings of Kilner *et al.* [8]. Assuming that the presence of carbides represents the main strengthening mechanism acting in the as-cast conditions, the diminution in the carbide content, associated with short solution treatments, is expected to have an adverse effect on the alloys' strength. The behaviour exhibited by the alloys with short solution treatment times indicates a significant contribution of work hardening, associated with the intrinsic stacking faults, to the alloys' strengthening.

It is known that the stacking fault energy is modified by compositional variations [3], being particularly sensitive to the amount of C dissolved in the matrix; the stacking fault energy increases with increasing C content. Thus, it is thought that this mechanism operates preferentially in the low C alloy, and that its participation is less important for the case of medium and high C alloys. In the former alloy the strength values were inferior to those corresponding to the latter alloys, although the intrinsic stacking faults had a more significant contribution to work hardening. This behaviour was associated to a relatively lower carbide content in the low C alloy. In contrast, in the medium and high C alloys, the carbide content was higher and thus its contribution to work hardening was more important than the effect of the intrinsic stacking faults. Thus, the resulting strength arouse from a combined effect of both mechanisms, i.e., the intrinsic stacking faults and the carbide content. It is noteworthy a marked difference existing between the strength levels of high and medium C alloys, the latter alloy exhibiting a higher value. In general, for treatment times of less than 60 min, for which an increase in the strength of both alloys was observed, the carbide content was higher for the medium C alloy. This suggests that in this case the carbides had a more important contribution to the alloy's strengthening. A more significant contribution of the intrinsic stacking faults to the alloy's strengthening is expected for the low C alloy, when compared with the high C alloy.

In the case of the low C alloy, the strength remained practically unalterable once a maximum value was reached. This indicated that the relatively slight variation existing in the carbide content beyond the maximum strength value was insufficient to promote posterior significant changes in the strength levels. In contrast, in the medium and high C alloys the strength was affected adversely after the maximum strength value was achieved, particularly in the case of the former alloy, which presented a behaviour opposite to that initially observed. This was attributed to a diminishing strengthening effect associated with a decrease in the carbide content, as well as to an insufficient contribution of the intrinsic stacking faults to compensate for the adverse effect of the carbide dissolution. In addition to that, as the solution treatment progressed, the C content of the matrix increased as a result of a redistribution of solute coming from the matrix/carbide interface, which increased the stacking fault energy and therefore diminished the contribution of the stacking faults to the alloy's strengthening.

4. Conclusions

ASTM F-75 alloys with various carbon contents were solution treated for several intervals of time. The tensile properties of the treated alloys were evaluated as a function of treatment time. According to the results obtained, the following conclusions can be drawn:

- The optical microscopy analysis showed that the as-cast microstructures were significantly sensitive to the carbon content, particularly with regard to the content and type of carbides formed. The as-cast microstructural features influenced the strength level of the alloys, while their ductility was unaffected.
- The solution treatment enhanced ductility, regardless of the alloy and duration of the treatment.
- The diminution in the carbide content caused by the solution treatment had different effects on the alloys' strength, depending on the duration of the treatment; short treatment times resulted in an increased strength, in contrast, the extensive carbide dissolution associated with long treatments had an adverse effect on the strength.
- For the solution treated alloys, the carbon content influenced markedly the resulting mechanical properties. The alloys with medium carbon content showed a satisfactory response to the solution treatment with respect to the mechanical properties specified by the ASTM F-75 standard.

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References

1. F. R. MORRAL, *J. Mater.* **1** (1966) 384.
2. K. ASGAR and F. A. PEYTON, *J. Dent. Res.* **40** (1961) 63.
3. C. P. SULLIVAN, J. D. VARIN and M. J. DONACHIE, in "Cobalt-Base Superalloys" (Centre d'Information du Cobalt, Brussels, 1970) p. 11.
4. M. RIDDIHOUGH, *Found. Trade. J.* **5** (1959) 421.
5. M. GÓMEZ, H. MANCHA, J. RODRÍGUEZ, J. ESCOBEDO, M. CASTRO and M. MÉNDEZ, *J. Biomed. Mater. Res.* **34** (1997) 157.
6. J. B. VANDER SANDE, J. R. COKE and J. WULFF, *Metall. Mater. Trans.* **A7** (1976) 389.
7. K. RAJAN and J. B. VANDER SANDE, *J. Mater. Sci.* **17** (1982) 769.
8. T. KILNER, W. M. LAANEMAE, R. PILHAR, G. C. WEATHERLY and S. R. MACEWEN, *ibid.* **21** (1986) 1349.
9. H. S. DOBBS and J. L. M. ROBERTSON, *J. Biomed. Mater. Res.* **18** (1983) 391.
10. T. KILNER, R. M. PILLIAR and G. C. WEATHERLY, *ibid.* **16** (1982) 63.
11. J. COHEN and R. M. ROSE, *ibid.* **12** (1978) 935.
12. C. O. MONTERO, M. TALAVERA and H. F. LÓPEZ, *Metall. Mater. Trans.* **A30** (1999) 614.
13. Z. GARZA, M. HERRERA, M. CASTRO, E. RAMÍREZ, M. MÉNDEZ and J. MÉNDEZ, *J. Mat. Eng. Perf.* **10** (2001) 153.
14. H. MANCHA, E. CARRANZA, I. ESCALANTE, G. MENDOZA, M. MÉNDEZ, F. CEPEDA and E. VALDÉS, *Metall. Mater. Trans.* **A32** (2001) 979.
15. M. CAUDILLO, M. HERRERA, M. CASTRO, E. RAMÍREZ, C. GONZALEZ and J. ISLAS, *J. Biomed. Mater. Res.* **59** (2002) 378.
16. E. RAMÍREZ, M. CASTRO, M. MÉNDEZ, J. LACAZE, M. HERRERA and G. LESOULT, *Scrip. Mater.* **47** (2002) 811.

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