

A Study on the Mechanical Properties of Dual Phase Steels Prepared by the Direct and Step Quenching Procedure

T. Dallali Isfahani ^{1*}, H. Sharifi ²

¹ Department of Materials Science and Engineering, Golpayegan University of Technology, Golpayegan, Iran

² Faculty of Engineering, Shahrekord University, Shahrekord, Iran

Abstract

Two different dual phase steels were prepared from low carbon manganese steel after the elimination of the banding using the direct quenching (DQ) and the step quenching (SQ) procedures. Different heat treatments resulted in different martensite morphologies, microstructures and mechanical properties. The heat treatments were designed in such a way to obtain a 0.25 volume fraction of martensite (V_m). For this reason, an intercritical temperature of 725 °C was applied. Furthermore, the tensile and impact properties were discussed. The results from the impact and tensile tests at different temperatures showed that the ductile-brittle transition temperatures for the DQ and SQ treatments were -49 and -6 °C respectively, while the DQ had better toughness.

Keywords: Dual phase steels, Fixed volume fraction of martensite, Impact properties, Toughness.

1. Introduction

Dual phase steels, or DP steels, refer to a class of high strength low alloy steels (HSLA) developed in the 1970s from conventional low alloy steels. They are composed of two phases; they are normally made of a soft and ductile ferrite matrix and a relatively hard dispersed phase of martensite, retained austenite and/or bainite depending on the cooling rate. The development was driven by the need for new high strength steels without reducing the formability or increasing the costs and supplying good elongation with a combination of strength and ductility. These steels are especially valuable for the automotive industry as it seeks formability, fatigue, crash resistance and weld ability, without influencing production cost required for ensuring the high tensile elongation, high tensile strength and low alloy content. Furthermore, the combination of high strength and good formability can reduce the weight of vehicles and provide environmental and economic advantages ¹⁻¹¹. In the last few decades, the microstructures, chemical composition, mechanical properties and the formability of these steels have been extensively studied ¹². Dual phase steels are generally produced by some intercritical annealing in the austenite + ferrite region followed by a rapid cooling to ensure the transformation of the austenite to the martensite. There are three basic approaches

for the commercial production of dual phase steels: a) the as-hot-rolled approach, where the dual-phase microstructure is developed during the conventional hot-rolling cycle by carefully controlling chemistry and processing conditions ¹³, b) the continuous annealing approach, where hot- or cold rolled steel strip is uncoiled and annealed intercritically to produce the desired microstructure ¹⁴, and c) the batch-annealing, where hot- or cold-rolled material is annealed in the coiled condition ¹⁵.

The formation of intercritically annealed dual phase steel is dependent on the parent/starting microstructure of the steel ¹⁶⁻¹⁸ and can be classified in terms of intermediate quenching, intercritical annealing (direct quenching), and step quenching. The nucleation and growth of dual phase steel can be different for each formation path. The “intermediate quenching” process starts with the production of martensite, which is then heated into the two-phase range. Austenite is nucleated at the martensite lath boundaries and grown accordingly. As carbon is diffused to the austenite, the remainder of the prior martensite is transformed into ferrite. The final microstructure is some kind of needle-like martensite dispersed in ferrite. The “intercritical annealing” process involves the annealing of ferrite and pearlite. When heated into the ($\alpha + \gamma$) region, austenite is nucleated at the interface between ferrite and carbide inside the pearlite colonies and grown into the carbide as the carbides are dissolved. The resultant microstructure, when the steel is quenched to room temperature, is some fine and globular martensite distributed along the ferrite boundaries. The starting microstructure of “step quenching” is austenite. When the austenite is cooled to a temperature between

* Corresponding author

Tell/Fax: + 9831 357240067

Email: t.isfahani@gut.ac.ir

Address: Department of Materials Science and Engineering, Golpayegan University of Technology, Golpayegan, Iran

1. Assistant Professor

2. Assistant Professor

Ar_1 and Ar_3 (critical phase transformation temperatures during cooling), ferrite is nucleated at the prior austenite boundaries and grown into the austenite. The subsequent structure after quenching is coarse martensite surrounded by ferrite. The grains obtained from “step quenching” are larger and coarser in comparison to those obtained from the other two formation processes.

Using dual phase steels with a 0.25 volume fraction of martensite makes the best combination of the useful properties in the automotive industry^{19, 20}. Some of the factors that influence the impact properties are morphology, distribution and carbon content of the martensite phase and also, the grain size²¹. Previous studies have mainly focused on high- martensite dual phase steels yielding low ductility and toughness properties. Furthermore, the studies on low-martensite dual phase steels were mainly concerned with fatigue properties. Due to the lack of information on the impact properties, the charpy curves are obtained and from these curves, the ductile- brittle transition temperature can be identified.

The main goal of the current study was to obtain the impact and tensile properties of dual phase steels with 0.25 V_m martensite, different morphologies and the distribution of martensite with a fixed grain size; also, the study was aimed at obtaining the Ductile-Brittle Transition Temperature (DBTT) for the different microstructures of dual phase steels with 0.25 V_m .

2. Experimental Procedure

The initial starting material used in this research was a hot rolled low carbon manganese steel plate of 10mm thickness with a ferrite-pearlite microstructure in the banding form (Fig. 1) and a composition shown in Table1. Before applying the heat treatment procedures for obtaining dual phase steels, the bonding structure had to be eliminated and for this reason, the initial steel was homogenized at 1200 °C for 4 hours and furnace cooled. Furthermore, to obtain a smaller grain size, the homogenized sample was normalized at 910 °C for 20 min. Finally, to obtain a dual phase structure, the heat treatment was applied between the Ac_1 and Ac_3 regions. The Ac_1 and Ac_3 regions were calculated using Eq. (1) & Eq. (2).¹ According to the calculations, Ac_1 and Ac_3 were 715 and 820 °C, respectively.

$$Ac_1 = 723 - 10.7Mn - 16.9Ni + 29.1Si + 16.9Cr + 290As + 6.38W \quad \text{Eq. (1)}$$

$$Ac_3 = 910 - 203 - 15.2Ni + 44.7Si + 104V + 31.5Mo + 13.1W \quad \text{Eq. (2)}$$

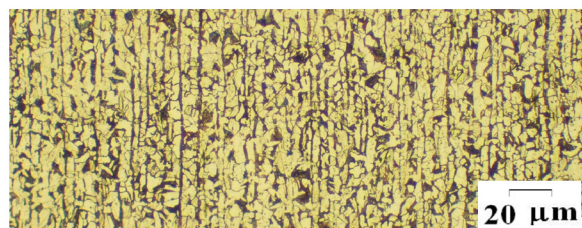


Fig. 1. Banding present in the ferrite-pearlite microstructure of the initial steel.

Table1. Composition of the steel used.

Steel Composition (wt. %)					
Fe	C	P	S	Si	Mn
Balance	0.137	0.028	0.022	0.30	1.57
Cr	Mo	Ni	Al	Cu	Nb
0.02	0.01>	<0.02	0.03	0.01	0.02

To obtain a 0.25 volume fraction of martensite, the intercritical thermal treatments were performed at 715, 725, 750, 775 and 800 °C for 1 hour and this was followed by quenching in brine solution. After the metallographic preparation of the samples, they were chemically etched with 3% nital. After etching, the microstructures of the samples were analyzed with the help of an optical microscope aided with Clemex image analysis software. This analysis was applied for the purpose of determining the martensite volume fraction and grain sizes. To obtain more accurate results, the image analysis was carried out in at least ten different regions of the specimen. As shown in Fig. 2, by both DQ and SQ heat treatment procedures, with an intercritical temperature at 725 °C, a 0.25 volume fraction of martensite was obtained.

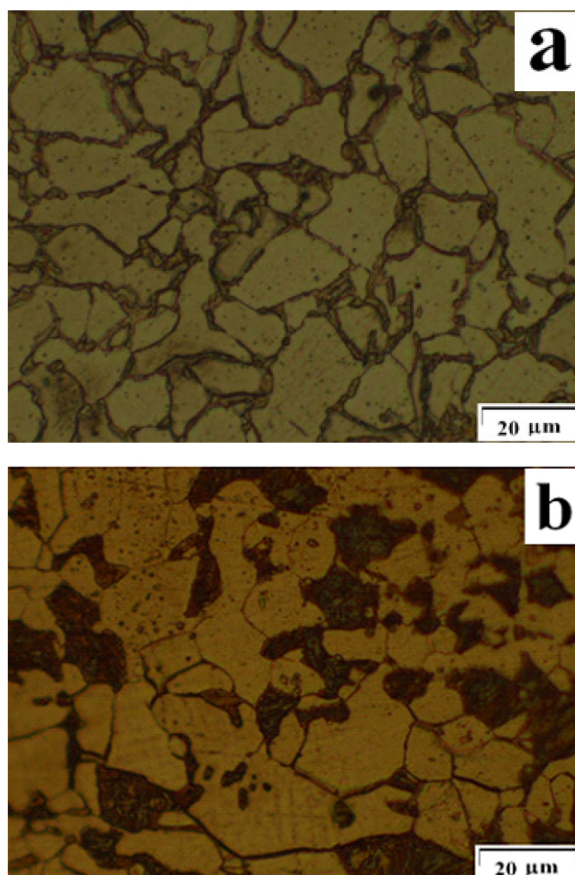


Fig. 2. Microstructure of dual phase steel achieved from (a) DQ ($\times 750$) and (b) SQ ($\times 750$).

The DQ and SQ heat treatment procedures are as Table 2. For the DQ procedure, the intercritical heat treatment was applied at 725 °C for 1 hour and this was followed by quenching in a brine solution which contained martensite with a spherical and network morphology while for the SQ procedure, the heat treatment consisted of austenitizing at 910 °C for 20 min, furnace cooling to the required intercritical temperature of 725 °C for 1 hour and finally, quenching in the brine solution. This procedure led to the formation of martensite in the aggregates of large, blocky shaped martensite islands. However, different heat treatment procedures are given in detail in Table 2. It should be noted that as shown in Table 2, in the previous heat treatment stage, the steels were homogenized and normalized at 1200 and 910 °C, respectively. Homogenizing was done to eliminate the banding while normalizing was applied to obtain smaller grain sizes and improve the mechanical properties.

Table 2. Heat treatment used to achieve a dual phase structure from the as received initial steel using direct quenching and step quenching procedure.

Method	Steps	Cooling method
Direct quenching	1- Homogenized at 1200 °C for 4h	Furnace
	2- Normalized at 910 °C for 20 min	Air
	3- Normalized at 910 °C for 20 min	Air
	4- Intercritical heat treatment at 725 °C for 1h	Water quenched (Brine solution)
Step quenching	1- Homogenized at 1200 °C for 4h	Furnace
	2- Normalized at 910 °C for 20 min	Air
	3- Normalized at 910 °C for 20 min and furnace cooled to 725 °C (intercritical heat treatment) and held for 1h	Water quenched (Brine solution)

As mentioned, a ferrite-martensite microstructure was expected to be obtained after applying the DQ and SQ heat treatment procedure. To ensure the formation of these phases, microhardness was applied. The results indicated the existence of a soft phase with a hardness value of around 200 Vickers while the hardness of the hard phase was obtained to be about 510 Vickers. These values indicated the formation of ferrite and martensite. At least six hardness values were performed at each sample. To obtain the tensile properties of the specimens, the tensile tests were conducted at room temperature after the specimen

preparation according to ASTM standard A370-B. The tensile tests were conducted using a computer controlled Housfield machine under a crosshead velocity of 0.50 mm/min.

3. Results and Discussion

Different heat treatment procedures applied were found to result in different morphologies of martensite which definitely varied the mechanical properties. The resulting true stress- true strain curves of the dual phase steels, as obtained by the DQ, the SQ, and the homogenized and austenitized initial steels, are shown in Fig. 3. In contrast to the initial steel with a non-continuous yielding characterized by luder bands (Fig. 3a), the resulting true stress- true strain curves of the DQ and SQ specimens showed a continuous yielding which was the unique characterization of ferrite- martensite dual phase steels (Figs. 3 b and c). As shown in Fig.3, the true stress- true strain curves indicated that the obtained dual phase steels had better properties compared to the initially homogenized and normalized steel. The reason for the better properties was that the hard martensite phase substituted the pearlite phase and acted as reinforcement in the dual phase steel. Also, the dual phase microstructures had continuous yielding which was a unique property of the dual phase steels due to their composite microstructure. For this reason, the dual phase steels had desirable formed surfaces while the initial steel had non-continuous yielding, resulting in bad surfaces formed. Compared to the dual phase steel obtained by the DQ procedure, the specimen from the SQ procedure had a better tensile property. It has been mentioned by Crawley et. al.²²⁾ and Sherman et. al.²³⁾ that due to the 3-4 volume percent expansion forced by the austenite to martensite transformation, the dislocation concentration around the brittle martensite phase is high, causing the ferrite phase to be under stress. In dual phase steels obtained by the DQ procedure, a uniform distribution of small spherical network martensite was obtained that provided a more uniform distribution of dislocations. However, the uniform distribution of dislocations resulted in less concentration of dislocations and also, less locking of dislocations. The SQ specimens contained large blocky martensite islands, resulting in a non-uniform distribution and concentration of martensite in different parts of the specimen. This was due to its less uniformity compared to the spherical and network morphologies of the DQ specimen that originated from the larger size of the blocky islands. This higher concentration resulted in more locked dislocations providing higher tensile properties. As shown by the true stress- true strain curves in Fig.3, the work hardening in the SQ specimen was higher than that in the DQ specimen due to the higher amount of locked dislocations. According to the mentioned

reasons, as shown by the curves, it could be understood that the DQ specimens had more elongation and better formability.

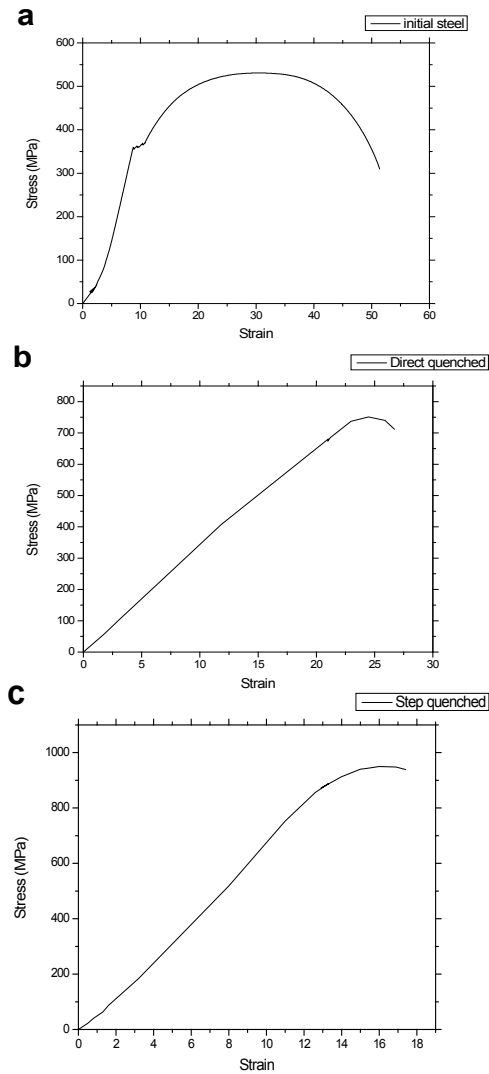


Fig. 3. The stress-strain curve of the: (a) homogenized and normalized initial steel, (b) specimen obtained by DQ and (c) specimen obtained by SQ procedure.

High work hardening rate and low yield stress to tensile strength ratio ($\frac{YS}{TS}$) of steels are other important characteristics of dual phase steels, making them suitable for many industries such as automobile industry. The $\frac{YS}{TS}$ ratio for the DQ specimen, SQ specimen and the initial steel was 0.54, 0.55 and 0.676, respectively. The other tensile properties are given in Table 3. The work hardening rate could be obtained by the slope of the curves.

Table 3. Tensile properties of the direct quenched, step quenched and initial steel.

	Direct quenched	Step quenched	Initial Steel
YS (MPa)	406	523	359
UTS (MPa)	751	950	531
$\left(\frac{YS}{TS}\right)$	0.540	0.550	0.676

Furthermore, impact tests were carried out on the V notch specimens by the Charpy method to obtain the ductile-brittle transition temperature. The Charpy curves of the initial steel (homogenized and normalized), DQ and SQ specimens containing two different microstructures of the dual phase steels were obtained as presented in Fig. 4a, b and c, respectively. According to the plotted Charpy curves from the data obtained by the impact test, the DBTT for the DQ, SQ and the initial steel were obtained to be -49, -6 and -34 °C, respectively. Therefore, it can be seen that the impact property of the DQ specimen was improved from the initial steel while the SQ specimen became the worst. The worst impact property of the SQ specimen was due to the large number of locked dislocations resulting in low ductility. However, at low temperatures, the mobility of dislocations was low, leading to more locked dislocations in different parts of the highly concentrated martensite. Due to all these reasons, lower impact energy and impact strength were obtained. Furthermore, in the DQ specimen, the more ductility resulted in better impact property. However, it can be seen that although the impact strength of the SQ specimen had a low value, that of the initial and DQ specimen was great.

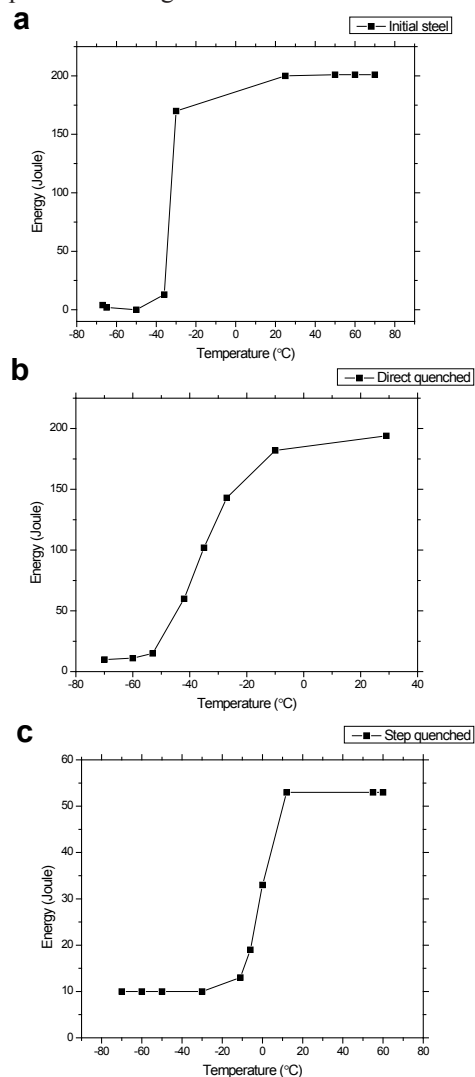


Fig. 4. Charpy curves of the: (a) homogenized and normalized initial steel, (b) DQ, (c) and SQ specimens.

The non-continuous formation of the martensite along the grain boundary in the SQ specimen resulted in poor impact property. Finally, the fractured surfaces of the impact specimens were analyzed using a scanning electron microscope (Figs. 5 and 6) for temperatures above and below the DBTT for dual phase steels prepared by the SQ and the DQ procedures. In the micrographs, the dimples were an indication of the ductile fracture while the cleavage fracture was reflective of the brittle fracture. From the SEM of the fractured surfaces shown in Fig. 5, as taken from the impact samples applied at temperatures higher than the DBTT, it could be understood that for the DQ specimens at $-47\text{ }^{\circ}\text{C}$, only dimples were present (Fig. 5a) while for the SQ specimen at $-5\text{ }^{\circ}\text{C}$, there were some areas with cleavage fracture along with areas of dimple fracture (Fig.5b). This indicated that the SQ specimens, compared to the DQ specimens, had a more brittle property. Furthermore, smaller dimples and a more uniform distribution can be seen from Fig. 5a (DQ specimen), as compared to Fig. 5b (SQ specimen). Also, from Fig. 6b (SQ specimen at $-30\text{ }^{\circ}\text{C}$), only cleavage fracture was seen while in the DQ specimen at $-50\text{ }^{\circ}\text{C}$ (Fig. 6a), both dimples and cleavage were seen. Furthermore, according to Fig. 6b, large non-uniformly distributed cleavage parts concentrated in different parts of the microstructure could be seen, probably due to the presence of large blocky martensite islands.

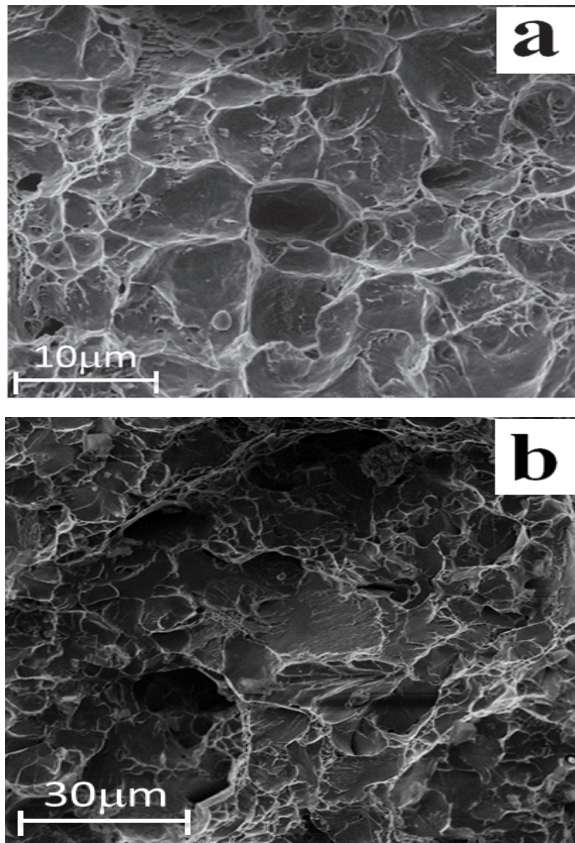


Fig 5. Fracture surfaces of the dual phase steel obtained from: (a) DQ and (b) SQ specimens above DBTT temperatures showing a ductile fracture at $-47\text{ }^{\circ}\text{C}$ and $-5\text{ }^{\circ}\text{C}$ respectively.

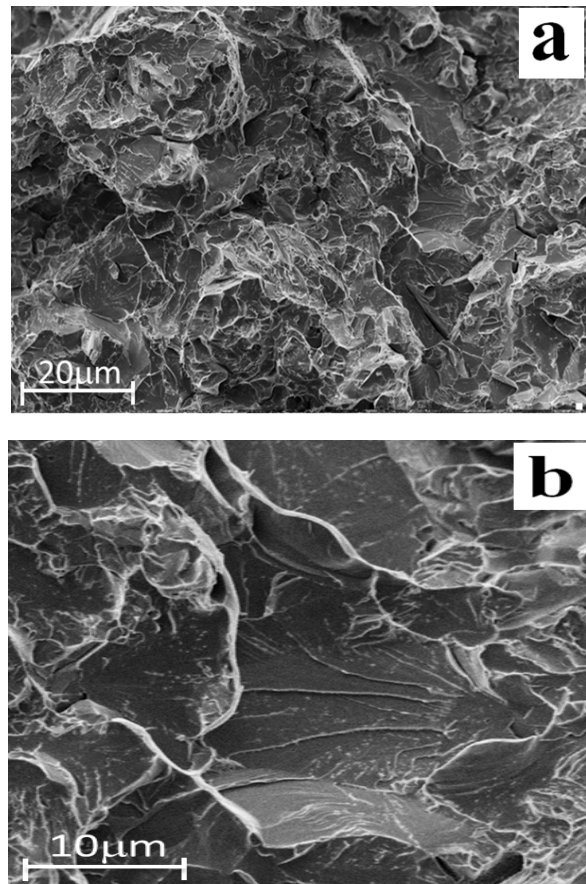


Fig. 6. Fracture surfaces of the dual phase steel obtained from: (a) DQ and (b) SQ specimens under DBTT temperatures showing a brittle fracture at $-50\text{ }^{\circ}\text{C}$ and $-30\text{ }^{\circ}\text{C}$ respectively.

4. Conclusion

As discussed, it can be concluded that both dual phase steels obtained by DQ and SQ procedure showed suitable tensile properties. It was found that the dual phase steel prepared by the DQ procedure had a better ductility and the dual phase steel prepared by the SQ procedure had better tensile strength. The YS was 406, 523 and 359 for the DQ, the SQ and the initial steel, respectively. On the other hand, for the UTS, they were 751, 950 and 531, respectively. However, according to the results, it can be concluded that the specimen prepared by the SQ procedure had such a bad impact property that made the use of this microstructure impossible and unsuitable. Furthermore, the SQ procedure deteriorated the impact property while DQ procedure improved it. The DBTT for the DQ and SQ specimens and the initial steel (homogenized and normalized) were -49 , -6 and -34 , respectively. However, as a final conclusion, it was shown that the dual phase steels obtained from the DQ procedure provided a desirable combination of impact and tensile properties, thereby showing to be a suitable candidate for several products.

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